Stellingen

behorende bij het proefschrift
"Strip casting of aluminium alloys- process and properties":

1. Het effect van warmtebehandelingen op de eigenschappen van een aluminium legering is groter dan het effect van stripgieten of DC gieten.

2. Zonder de praktijk te kennen kunnen de resultaten van computersimulaties niet geïnterpreteerd worden.

3. Gepolijste huidvelden op de staart of vleugel van een vliegtuig maken het vliegtuig onnodig duur zonder dat het vlieggedrag beïnvloed wordt.

4. Het automatisch en direct opnemen van de telefoonhoorn wanneer de telefoon gaat, is een duidelijk voorbeeld van een Pavlov reactie.

5. Een relatie is vergelijkbaar met zand in je hand: hoe hardere je het probeert vast te houden des te eerder glipt het tussen je vingers weg.

6. Een vis is zich als laatste bewust dat hij in een viskom zwemt.

7. Roze en blauwe muisjes bij de geboorte van een kind is het begin van een stereotype rollenpatroon.

8. Peuterspeelzalen en crèches moeten maar één doel nastreven en dat is kinderen laten spelen, in plaats van kinderen (cognitief) voor te bereiden op de basisschoolperiode.
Theses

belonging to the PhD thesis:

"Strip casting of aluminium alloys - process and properties":

1. The effect of heat treatments on the properties of aluminium alloys is larger than the effect of strip casting or DC casting.

2. Results of computer simulations can not be interpreted without practical knowledge.

3. Polish skins on the tail or wing of an earoplane are unnecessarily increasing the costs without influencing the flying behaviour.

4. Taking automatically and directly the telephone horn when it rings is a clear example of a Pavlov reaction.

5. A relation can be compared to sand in your hand: the harder you try to hold it, the faster it will be slipping through your fingers.

6. A fish will be the last one to know that it is swimming in a fish bowl.

7. Pink and blue aniseed comfits when a child is born, is the start of a standard sex stereotyping.

8. Playgroups and nurseries should aim at one goal and that is to let children play instead of get the children (cognitive) ready for primary school.
Strip Casting of Aluminium Alloys
Process and Properties

Stripgieten van Aluminium Legeringen
Proces en Eigenschappen

E.N. Straatsma
Strip Casting of Aluminium Alloys
*Process and Properties*

Stripgieten van Aluminium Legeringen
*Proces en Eigenschappen*

Proefschrift

ter verkrijging van de graad van doctor
aan de Technische Universiteit Delft,
op gezag van de Rector Magnificus prof. dr. ir. J.T. Fokkema,
voorzitter van het College voor Promoties,
in het openbaar te verdedigen op maandag 14 oktober 2002 om 13.30 uur

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1 Introduction

Aluminium is nowadays a commonly used material due to its low weight combined with excellent strength, corrosion properties, and ductility [1, 2]. During the upcoming usage of aluminium it had to compete with the established materials such as wood, copper, and steel. In the late 19th century aluminium won its position and the industrial production began, therefore aluminium is often called a new material. The total consumption of aluminium is still growing and the largest market of aluminium is the transportation sector with 28% of the total consumption in 1995. Besides, the largest growth was found in the packaging market where the consumption increased with 75% from 1970 till 1995 [1].

This thesis is concentrating on thin strip casting of aluminium alloys and in particular AlMnMg alloys [3-5]. This alloy is extensively used in the beverage can industry for its good ductility combined with a good strength [6]. The strength of this alloy is obtained by strain hardening whereas additions of magnesium give some dispersion hardening.

1.1 The AlMnMg alloy

The liquidus projection and an isothermal section at 400 °C of the ternary AlMnMg alloy phase diagram [7] are given in Figure 1.1. At weight percentages Mn and Mg of 1%, only liquid Al is present at liquidus temperature. The maximum solubility of manganese in aluminium is 1.8 % at a temperature of 657 °C (930 K) when the alloy solidifies under equilibrium condition. Additions of Mg decreases the solubility of Mn. The solubility of Mn in aluminium can be increased to 15 % when the alloy is rapidly cooled. Manganese in solution attributes to the strength of the alloy.
Strip Casting of Aluminium Alloys

Figure 1.1. Liquidus projection (a) and an isothermal section at 400 °C (b) of the ternary AlMnMg alloy phase diagram. Data taken from [7].

The manganese forms with aluminium the intermetallic Al₄Mn as is seen in Figure 1.1b. This intermetallic is present as plate, needle or Chinese script and it decreases the ductility. The increase of strength can be accompanied by an increase of ductility due to a change of the microstructure. This is the case when the manganese intermetallics are rounded or in the form of Chinese script which does not embrittle the alloy. Additions of magnesium or impurities such as iron and silicon are responsible for the formation of Al(FeMn), Al(FeMn)Si, AlMg and Mg₂Si intermetallics. The presence of these intermetallics depends on alloying content and solidification rate. The composition limits of an AlMnMg alloy such as AA3004, specified by the Aluminum Association, is given in Table 1.1.

Table 1.1. Composition limits of AA3004.

<table>
<thead>
<tr>
<th>Element</th>
<th>AA3004 (weight %)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mn</td>
<td>1-1.5</td>
</tr>
<tr>
<td>Mg</td>
<td>0.8-1.3</td>
</tr>
<tr>
<td>Fe</td>
<td>≤ 0.70</td>
</tr>
<tr>
<td>Si</td>
<td>≤ 0.30</td>
</tr>
<tr>
<td>Cu</td>
<td>≤ 0.25</td>
</tr>
<tr>
<td>Zn</td>
<td>≤ 0.25</td>
</tr>
<tr>
<td>Ti</td>
<td>≤ 0.15</td>
</tr>
<tr>
<td>Others</td>
<td>≤ 0.15</td>
</tr>
</tbody>
</table>
1.2 Rapid solidification of aluminium alloys

Solidification of aluminium [8] occurs when the melt cools down from its melting temperature, which is 658 °C for AA3004, to a critical temperature where nuclei are formed. This critical temperature is a few degrees below the liquidus temperature and is called the undercooling of the melt. The nuclei grow out to grains and latent heat is released that increases the temperature of the melt. Since the latent heat is flowing to an area of lower temperature the melt finally solidifies.

In the case of single-roll strip casting the nuclei needed for solidification are formed in a heterogeneous way due to the chilled surface. Rapid solidification occurs and results in a thin layer consisting of several grain structures. Three layers can be distinguished from the chilled surface to the centre of the casting: fine, columnar, and coarse equiaxed grains respectively. The thickness of each layer depends on composition and cooling conditions.

Rapid solidification reduces the length of the columnar zone and gives smaller grains. Smaller grains increase the strength and formability properties. In addition, due to high cooling rates more alloying elements remain in solid solution which affect the properties of the material. This aspect challenges to develop dedicated alloys which are not possible with conventionally process routes.

1.3 Strip casting

Strip casting [1, 9, 10] is a rapid solidification process and refers to the production of continuously cast strip or sheet. There are a number of variants of this process [10] with the common aspects that strip is formed on (or between) a moving surface. The concept of casting liquid steel between two cooling rolls is developed and patented in 1846 by Sir Henry Bessemer [11]. From this date, alternatives are developed with surfaces as belts, blocks, or a combination with rolls. Figure 1.2 shows some types of casters which are based on the patent of 1846 and which are used on industrial scale. These casters, which produce wide strip between 1000 and 2000 mm, can be divided into thick and thin strip casters. The twin-belt casters and block casters produce thick strip, typically 10 to 20 mm thick, whereas twin-roll casters produce thin strip, typically 3 to 10 mm thick.
Strip Casting of Aluminium Alloys

Figure 1.2. Strip casters. 1) twin roll caster (Pechiney-Coquillard); 2) belt caster (Hazelett); 3) block caster (Hunter-Douglas). The arrow indicates the liquid metal flow.

The twin-belt Hazelett caster [12] is manufactured in the USA by Hazelett Company and is used for casting slabs of zinc, aluminium alloys, copper, and copper alloys. The slabs are directly hot rolled in line to a thickness of 1-3 mm. The aluminium sheet is used for can stock and automotive body panel sheet. The block casters, such as the Hunter-Douglas, Alusuisse Caster II, and the Lauener block caster, are used to cast aluminium alloys. The cast slab is hot rolled in line [1, 10]. The twin-roll casters are used for casting alloys with a short solidification range such as pure aluminium, the 3xxx series, and the low magnesium containing 5xxx series. The diameter of the roll influences the casting rate which is roughly 0.1 m/s with gauges of 4-6 mm. Worldwide there are around 100 casters of this type [1]. The advantage of two cooling surfaces is that both surfaces of the strip are smooth and of good quality. As a consequence and disadvantage, the microstructure shows five layers of which the centre contains all the impurities. This centreline segregation causes quality problems [13].

Single-roll casters are used to produce small strip or ribbon at high wheel speeds. Due to these high speeds, the thickness can be in the order of microns although the strip is limited in width (20-50 mm). This process concept is used in planar flow casting and melt spinning, producing strip and ribbon respectively [14, 15].

With the former casters it is difficult to produce wide strip (sheet) in combination with a thickness in the range of 1-3 mm without quality problems. To study the effect of alloying elements and process parameters on the properties of thin sheet with a thickness in the order of 1-3 mm, a laboratory single-roll strip caster is used. With a single-roll strip caster it is possible to cast alloys with a wider solidification range and without centre-line segregation. It has to be remarked that it is not our intention to develop a strip caster.
A requirement to study the properties of the strip is to obtain a good quality of the free surface. The surface and the strip quality is influenced by the way it is poured
onto the wheel. In single-roll strip casting two types of feeding systems can be used, a liquid drag system and a vertical feeding system [10]. The liquid drag system contains a feeding system with liquid metal that is positioned against the cooling surface. The cooling surface builds up a solid layer by rotating slowly through the molten metal. The thermal conditions inside the feeding system are hard to control since the local temperature is not known. To avoid these problems a vertical feeding system is chosen in this study. This vertical feeding system is positioned on top of the rotating wheel and supplies molten metal through an rectangular slit onto the wheel.

1.4 Production of sheet

Strip casting results in strip with gauges of 1 to 25 mm and is therefore an economically interesting process for producing sheet. In Figure 1.3 the strip casting process is compared to the Direct Chill (DC) casting process, which is the conventional way of producing sheet material. Conventionally produced sheet is supplied to the can stock industry, automotive industry, and aerospace industry for the final processing such as forming operations. Prior to the forming operation, the sheet is mostly heat treated to increase the ductility and to remove residual stresses caused by cold rolling.

The conventional process route of an AlMnMg alloy for achieving a thickness of approximately 0.3 mm takes place as follows [16]. A DC cast slab of 300-760 mm thick and 2000 mm wide is first scalped and subsequently homogenised at temperatures in the order of 550 °C. The slab is then cooled to approximately 500 °C and hot rolled in two stages. The first stage takes place in a breakdown mill to obtain a sheet thickness of 25 mm. The temperature of the sheet is decreased to approximately 300 °C, where in the second stage, the sheet is reduced to a thickness of 2.5 mm in a finishing mill. Finally the sheet is cold rolled to a typical thickness of ≤ 0.3 mm. In contrast, strip casting is directly cold rolled to a final strip thickness of 0.3 mm and, depending on the casting conditions, it is not needed to homogenise prior to rolling. By eliminating the hot rolling steps, strip casting is an energy saving process.
1.5 This Thesis

This thesis studies the properties of directly cast thin aluminium strip as a function of process conditions. The single-roll strip caster is therefore used as a tool to produce thin strip and to cast different alloys. Figure 1.4 shows schematically the content of this thesis.
Figure 1.4. Schematic representation of the strip caster and the content of this thesis.

Chapters 2 and 3 are dealing with the single-roll strip casting process whereas chapters 4 till 6 are dealing with the properties of the strip cast material. Chapter 2 describes the flow behaviour in the feeding system. In this chapter the influence of the geometry of the feeding system on the flow behaviour is studied by experiments and by simulating the flow behaviour in the feeding system three dimensionally. Chapter 3 describes the formation of thin strip onto the wheel. In this chapter the thickness of the strip as a function of process parameters is studied by simulating the strip formation two dimensionally. Chapter 4 describes the as-cast microstructure of the strip. In this chapter the microstructure is studied as a function of process parameters. Chapter 5 describes the microstructure and the tensile properties of cold rolled strip. In this chapter the rolled microstructure is studied as a function of heat treatments and alloy composition. Chapter 6 describes the formability characteristics of the rolled and heat treated material. In this chapter the texture and anisotropy of rolled strip is studied as a function of heat treatments and alloy composition. In addition the strip cast material is compared to conventionally produced material.
Strip Casting of Aluminium Alloys

References
2 Feeding system of a single-roll strip caster

2.1 Introduction

Aluminium sheet or strip is produced nowadays in two ways, namely via direct chill casting [1] or via strip casting [2]. The conventional and most common way of sheet production is by hot rolling and cold rolling of DC-cast ingots, usually of rectangular cross-section. Prior to hot rolling the ingots are homogenized to remove any casting inhomogeneities followed by scalping off the shell zone and oxide layer. In strip casting sheet is produced directly from the melt, eliminating the need for homogenizing, scalping and hot rolling [3]. Strip casters can be divided into thick strip casters and thin strip casters. Thick strip casters produce strip with a thickness of 25 to 30 mm and the sheet properties correspond to those of conventionally produced sheet. Thin strip casters used in industry produce strip with a thickness of 3 to 12 mm and the properties of the strip are different since the cooling rate is appreciably greater than in the case of DC casting. This high cooling rate gives opportunities for specific material design [1].

A critical aspect of strip cast process is the feeding system, which should provide a stable melt flow on the wheel without any turbulence. Pouring into a feeding system will always give inclusions and oxide films that will influence the flow. Also thermal aspects such as natural and forced convection inside the feeding system determine the outlet flow. The outlet flow is of great importance to the solidification behaviour of the strip. The outlet flow determines the contact with the wheel and affects therefore the heat transfer and solidification rate. Since the solidification behaviour affects the performance of the strip, the fluid flow behaviour in feeding systems has been subject to extensively research [4-12]. A review is given in [8]. This research is generally concentrating on determining the time a fluid element spent in the feeding system, also called the residence time. The spread in residence times should be minimal, because this indicates an equal velocity throughout the feeding system. The residence time is influenced by flow controls, such as baffles and dams. The performance of these flow modifiers depends on the method of pouring and the design of the feeding system. Computer simulations showed that temperature fields inside a feeding system are not much affected by flow modifiers [10].
Strip Casting of Aluminium Alloys

Sato et al [11] and Jefferies et al [12] studied the flow behaviour of a feeding system with a porous plug in front of the exit slit as a flow modifier. They found that the porous medium gave a uniform distribution of the fluid through the exit slit and improved therefore the outlet flow. Numerical simulations [12] proved that it smooths the flow because the velocities are more uniform and that it prevents the formation of dead zones at the back of the nozzle wall. Dead zones lead to accumulation of solidified metal and result in freezing off the tip. The porous medium improves the surface quality of the strip [11].

The aim of this study is first to describe the laboratory single-roll strip caster that is developed for the experiments. Secondly, experiments are reported which are performed to optimise the outlet flow of the feeding system and finally, the fluid flow is modelled inside the feeding system.

2.2 Configuration of the single-roll strip caster

The single-roll strip caster able to produce aluminium strip with gauges of 0.7 to 3 mm and widths up to 100 mm is shown in Figure 2.1. The caster consists of a frame with a rotating wheel, which has a diameter of 300 mm.

![Figure 2.1. Single-roll strip caster. A is the casting roll; B is the water supply coming from a reservoir, C is the feeding system.](image-url)
The rotating wheel, which is the casting roll, consists of a CuCr tube with a length of 200 mm which is fixed on a stainless steel axle. The CuCr tube has a wall thickness of 50 mm and, inside, a screw profile is machined to provide optimum water-cooling. The water enters the tube via the stainless steel axle by a neutral bearing coupling and the water exits at the other side of the axle where it flows into a reservoir of 100 litres. The reservoir, which is part of a closed circuit, is equipped with a thermostat to establish the water temperature that is adjustable to 90 °C. The roll is driven by an electric motor via a single chain and the roll speed is adjustable up to 500 RPM.

![Cross-section of the feeding system](image)

**Figure 2.2. Cross-section of the feeding system.** A is the pouring tap, which provides a constant metallostatic pressure; B is the flow distributor; C₁ is the top of the distribution chamber, C₂ is the bottom of the distribution chamber, D is the exit slit; E is the distance between wheel and flow distributor (= gap distance). Dimensions are in millimetres.

Liquid metal is cast in a vertical feeding system (Figure 2.2) that is positioned with a holder on top of the casting roll. The feeding system consists of two parts, a pouring tap and a flow distributor, indicated in Figure 2.2 by a rough hatched area and a fine hatched area respectively. The bottom of the flow distributor rests on top of the wheel and therefore it is necessary to machine a slit into the bottom as wide as the exit slit. This slit becomes the gap between wheel and flow distributor. The liquid metal inside the pouring tap provides a metallostatic pressure.

The flow from the exit slit to the wheel is hindered by the first splashing on the wheel and by the formation of a *vena contracta* [13]. A *vena contracta* occurs when the direction of the moving liquid suddenly changes. The gap distance should be small to suppress the splashing and the flow on to the wheel should be gradual to minimise the *vena contracta*. Further details of the configuration of the feeding system are given in section 2.3.1.
2.3 Experiments

2.3.1 Configuration of the feeding systems

The feeding systems are made of an insulating material, Marinite®, and consist of a wedge shaped pouring tap and a flow distributor (see Figure 2.2). The parts are divided in the centre, which makes it easy to unload the system and to machine it from plates. The thickness of the exit slit was 0.5 mm or 1.5 mm. The distribution chamber has a height of 18 mm and vertical holes were machined in it. The diameter and the number of vertical holes in this chamber were varied. Table 2.1 and Figure 2.3 show the considered feeding systems.

<table>
<thead>
<tr>
<th>System</th>
<th>Number of holes</th>
<th>Tube diameter (mm)</th>
<th>Exit slit thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>2</td>
<td>3</td>
<td>0.5</td>
</tr>
<tr>
<td>2</td>
<td>7</td>
<td>3</td>
<td>0.5</td>
</tr>
<tr>
<td>3</td>
<td>13</td>
<td>3</td>
<td>0.5</td>
</tr>
<tr>
<td>4</td>
<td>7</td>
<td>6</td>
<td>1.5</td>
</tr>
<tr>
<td>5</td>
<td>continuous (α)</td>
<td>6</td>
<td>1.5</td>
</tr>
<tr>
<td>6</td>
<td>continuous (α)</td>
<td>6</td>
<td>0.5</td>
</tr>
</tbody>
</table>

Table 2.1. Configuration of the feeding systems

Figure 2.3. Part of the flow distributor with cross sections of the distribution chambers of the six feeding systems (Table 2.1).
Experiments were performed with four feeding systems, system 1 to system 4. The experiments were aimed to optimise the flow distributor of the feeding system in order to create a constant uniform flow as wide as the exit slit.

2.3.2 Results
The first system (2/Ø3/0.5mm) initially showed the formation of many ribbons (about 2 mm in diameter) such as seen in Figure 2.4a, followed by the formation of two strips of about 20 mm width each. Then, strip formation stopped because the fluid solidified in the exit slit. The solidified residual shells from the flow distributor showed a clear flow pattern with vortices and also showed that the pouring stream flowed back into the chamber when it entered the narrow exit slit.

The second system (7/Ø3/0.5mm) gave for a short time one strip as wide as the exit slit but it converged towards the centre of the exit slit resulting in one strip of about 35 mm wide. The solidified residue at the exit slit showed solidified droplets between the locations of the holes, which indicate that the exit slit is mainly fed by the holes. Clearly the number of holes is insufficient for complete filling of the exit slit.

![Figure 2.4. Strip that resulted from system 1 (a) and strip that resulted from system 4(b).](image)

The third system (13/Ø3/0.5mm) produced strip only at one side of the exit slit. The residue shell showed no waves inside the distribution chamber such as was seen for system 1. Apparently the liquid was distributed well in that part. The observation that the strip width was half of the exit slit width indicates that the fluid flow in the feeding system is sensitive to little disturbances caused by the small dimension of the exit slit.

The fourth system (7/Ø6/1.5mm) resulted in strip as wide as the exit slit, see Figure 2.4b. The feeding system was entirely emptied and therefore there was no
solidified residue. Apparently, this system leads to a well distributed liquid (and is less sensitive to disturbances than the previous systems). The thickness of the strip is about 1 mm and the surface quality of the strip is reasonably good. There are no pores. The surface roughness $R_a$ of the strip is smoother on the wheel side than on the other side, that is 1.2 μm and 4.3 μm, respectively.

Apparently there are two aspects that influence the outlet flow. First the dimension of the holes and exit slit and secondly the number of holes in the distribution chamber. By increasing the thickness of the exit slit, the flow becomes more reliable.

A uniform outlet flow results from a sufficient feeding of the exit slit which depends on the number of holes in the distribution chamber and in this way the holes act as a flow modifier.

2.4 Modelling

2.4.1 Modelling of the feeding system
In practice the feeding system is a black box and differences in the dimensions of the system result in different strips. An analytical approach is used to calculate mean velocities of the liquid inside the feeding systems but to gain more understanding about the fluid behaviour, the feeding systems are simulated. The goal of the simulations is to determine the effect of a flow modifier and to find an explanation for the differences that are found in practice between the systems.

2.4.2 Bernoulli
An analytical approach is the Bernoulli equation, written in terms of unit mass of material flowing:

$$\int_{z_1}^{z_2} \frac{dP}{\rho} + \left[ \frac{v_2^2}{2\beta_2} - \frac{v_1^2}{2\beta_1} \right] + g(z_2 - z_1) + M + E_f = 0 \quad (2.1)$$

where $P$ is the pressure, the indices 1 and 2 indicate in this case (Figure 2.5) the top and bottom of the column respectively, $\rho$ is the density, $v$ is the average velocity at the specific position, $\beta$ is a factor that depends on laminar ($\beta=0.5$) or turbulent ($\beta=1$) flow, $g(z_2-z_1)$ is the potential energy, $M$ is the mechanical work done by the system which is zero, and $E_f$ is the friction loss. There is no pressure difference
between inlet and outlet due to the open system and the velocity in the pouring tap is negligibly small compared to the exit velocity, hence the equation becomes:

\[
\frac{v_2^2}{2\beta_2} - g\bar{z}_1 + E_f = 0
\]  

(2.2)

The total friction loss is the sum of all resistances experienced by the fluid. The friction loss is determined by contractions and expansions in flow area and by the resistance of the inner walls. The friction loss due to contractions and expansions is determined as follows:

\[
E_f = \frac{1}{2} v^2 e_f
\]  

(2.3)

Hence, \(\bar{v}\) is the average velocity in the smaller area. The friction-loss factor \(e_f\) depends on the geometric ratio of the system and the Reynolds (\(Re\)) number and is evaluated graphically using the procedure that is described in [15].

The friction loss due to inner walls is evaluated with a friction factor \(f\) which is a function of the \(Re\) number, the length of the flow area \(L\), and the equivalent diameter \(D_e\):

\[
E_f = 2\bar{v}^2 \frac{L}{D_e} f = 2\bar{v}^2 \frac{L}{D_e} \left(0.0791 Re^{-1/4}\right)
\]  

(2.4)

The correlation for the friction factor \(f\) exists for \(2100 < Re < 10^5\). The equivalent diameter \(D_e\) is used for non circular shapes:

\[
D_e = \frac{4x \text{ flow area}}{\text{wetted perimeter}}
\]  

(2.5)
2.4.3 Computational fluid dynamics (CFD)

The filling of the feeding system is simulated with a commercially available CFD programme FLOW-3D®. This programme is used to investigate the dynamic behaviour of liquids including incompressible fluid flow and free fluid surfaces. The program is based on the laws of mass, momentum and energy conservation. The fluid interfaces and free surfaces are defined in terms of a Volume Of Fluid function (VOF) and satisfy the continuity equation.

A Finite Volume Method is used to solve the equations numerically, which means that the physical laws are satisfied over finite regions rather than at one point (i.e. Finite Difference Method). The flow region is subdivided into a fixed (Eulerian) grid of variable-sized rectangular cells. General geometric regions are defined with a free gridding technique. This means that the grids do not deform the geometry of obstacles. All variables are located at the centres of the cells except for velocities, which are located at the cell faces. Most terms in the equations are evaluated explicitly except the pressure forces. Pressures and velocities are coupled implicitly by using time-advanced pressures in the momentum equation and time-advanced velocities in the continuity equation. This semi-implicit formulation is accurate in case of low speeds and incompressible flows.

Three feeding systems are simulated, system 4 to system 6 (see Table 2.1 and Figure 2.2). The simulation is three dimensional and includes heat flow to the moving substrate. All feeding systems start with a complete filled area with liquid at the top boundary. The boundary condition at the top is located at the top of the flow...
Feeding system of a single-roll strip casterer

distributor (location B, see Figure 2.2) and represents a hydrostatic pressure of 2.3 kPa corresponding with a columnar height of 100 mm. The substrate is assumed flat and has a linear velocity of 0.63 m/s. Solid is moving with the same velocity as the substrate. During filling of the feeding system the assumption is made that the fluid is incompressible and that the flow is laminar (Re<3*10^5). Further, we assume that there are no heat losses between liquid and feeding system. The simulations are applied both with and without surface tension [14]. Figure 2.5 shows the coordinate system of the feeding system. The width of the feeding system corresponds with the width of the strip (0.05 m) and is presented as $x$. The thickness of the pouring tap is presented as $y$, and the height of the feeding system is $z$. The cells are smaller at critical areas and thus have different dimensions ($x, y, z$). The exit slit consists of 6 cells in the $z$ direction and a minimum of 3 cells in the $y$ direction. The gap distance (E in Figure 2.2) is taken equal to the exit slit thickness. The material properties are taken from the binary AlMn1 alloy. These properties and the process parameters are listed in Table 2.2.

<table>
<thead>
<tr>
<th>Table 2.2. Material properties of AlMn1 and process parameters [14, 15].</th>
</tr>
</thead>
<tbody>
<tr>
<td>liquidus temperature</td>
</tr>
<tr>
<td>solidus temperature</td>
</tr>
<tr>
<td>viscosity at melting point</td>
</tr>
<tr>
<td>surface tension coefficient</td>
</tr>
<tr>
<td>contact angle</td>
</tr>
<tr>
<td>latent heat</td>
</tr>
<tr>
<td>heat transfer coefficient</td>
</tr>
<tr>
<td>linear wheel velocity</td>
</tr>
<tr>
<td>specific heat</td>
</tr>
<tr>
<td>thermal conductivity</td>
</tr>
<tr>
<td>density</td>
</tr>
</tbody>
</table>

Figure 2.6 shows schematically the input, output, and variables used in the simulation.
Strip Casting of Aluminium Alloys

Variables:
- geometry feeding system
- surface tension

Input parameters:
- fluid behaviour
- solidification properties
- alloy properties
- wheel velocity
- hydrostatic pressure

Output parameters:
- flow rate at boundary
- velocity (x, y, z)

Figure 2.6. Schematic presentation of the simulation.

2.4.4 Results
1. Flow rate
The average flow rate \((Q = \nu x A)\) obtained from simulations of Flow3D\textsuperscript{®} in the steady state condition is equal for systems 4 \((7/\Ø 6/1.5 \text{mm})\) and system 5 \((\% /\Ø 6/1.5 \text{mm})\), which is \(0.41 \times 10^{-4} \text{ m}^3/\text{s}\). The flow rate of system 6 \((\% /\Ø 6/0.5 \text{mm})\) is much smaller since the outlet area is smaller. This indicates that in the simulation the flow rate is mainly determined by the columnar height of the melt and the outlet area, which are identical for the two systems, and not by the distribution chambers which are different. The value of the flow rate from the simulation is given in Table 2.3 together with the experimental flow rate, calculated from the thickness of the strip and wheel velocity. It is seen that the experimental flow rate and simulated flow rate are almost similar.

Table 2.3. Flow rates that are obtained by experiments, simulation (systems 4 \((7/\Ø 6/1.5 \text{mm})\) and 5 \((\% /\Ø 6/1.5 \text{mm})\)) and calculation. Assumption: exit slit thickness of 1.5 mm, columnar height of 100 mm, friction losses from walls, contraction and expansion.

<table>
<thead>
<tr>
<th></th>
<th>Average flow rate *10^{-4} m^3/s</th>
</tr>
</thead>
<tbody>
<tr>
<td>Experimental</td>
<td>0.32 ± 0.05</td>
</tr>
<tr>
<td>Flow3D\textsuperscript{®}</td>
<td>0.41 ± 0.01</td>
</tr>
<tr>
<td>Bernoulli laminar</td>
<td>0.74</td>
</tr>
<tr>
<td>Bernoulli laminar with friction losses</td>
<td>0.64 ± 0.01</td>
</tr>
<tr>
<td>Bernoulli turbulent</td>
<td>1.05</td>
</tr>
<tr>
<td>Bernoulli turbulent with friction losses</td>
<td>0.91 ± 0.02</td>
</tr>
</tbody>
</table>
The flow rates calculated from Equation 2.1 for laminar or turbulent conditions and considering possible friction losses are also given in Table 2.3. Friction losses are discussed in Section 2.5.2 and the most important contribution to those losses is the expansion of fluid at the exit area. It is seen that this equation overestimates the flow rate. Apparently, the friction losses should be taken larger.

2. Flow in the holes
The velocity increases when the fluid flows through the holes due to gravitational acceleration and an increase of columnar height. The velocity profile of a hole under steady state conditions is shown in Figure 2.7 at two locations, on top of a hole (C1, see Figure 2.2) and at the bottom of a hole (C2, see Figure 2.2). At the wall the velocity is zero (assumption of wall resistance) and increases to a maximum in the centre of the hole. The shape of the profile becomes sharper when the fluid flows through the holes because of the development of a velocity boundary layer.

![Figure 2.7](image)

*Figure 2.7. The z-velocity inside a hole (steady state) in the z-y plane, x=0. Location C1 --- and C2 ——.*

3. Recirculations inside the system
From the z-velocity inside the distribution chamber the Reynolds (Re) number is calculated. For the system with a flow modifier the Re number inside the distribution chamber is 3800 and without a flow modifier (system 5) the Re number is 4600. If the distribution chamber is seen as a closed conduit, the Re number for laminar-to-turbulent transition is 2100 [15]. Therefore the flow is turbulent inside the distribution chambers for both systems, as indicated by the Re number. Figure 2.8 shows vector plots in the x-y plane in the distribution chamber of system 4 (7/Ø6/1.5mm) and system 5 (Ø/Ø6/1.5mm) at two locations during steady state, when the fluid enters the holes (C1) and when it leaves the holes (C2). System 4
shows at the wall of the holes symmetric recirculations when the fluid enters the holes but these disappear when the fluid leaves the holes. This difference is seen in system 4 but not in system 5 and 6 where no flow modifier is applied. System 5 (Figure 2.8b) shows an uncontrolled flow at location $C_1$ that develops into a vortex and a large area of zero velocity (dead zone) at location $C_2$.

Figure 2.8. Part of a vector plot in the x-y plane. System 4 (a) and system 5 (b) at locations $C_1$ and $C_2$.

4. Filling patterns of the flow distributors
Since some of the experiments stopped in an early stage, also the time to steady state in the simulations are analysed. All feeding systems start with a complete filled area with liquid at the top boundary. In practice, the systems are filled with a crucible which means a stream of liquid fills the system. This stream is always aimed at the tapered wall (A in Figure 2.2) to promote a wide flow and to prevent the liquid to flow directly into the exit slit. The initial filling behaviour of the systems that are simulated with a completely filled area at the top boundary, comes close to the experimental situation because the fluid looses its horizontal shape and becomes a stream, as can be seen in Figure 2.9. Simulations which are performed with a stream of diameter 10 mm showed a more turbulent filling behaviour of the distribution chamber but it resulted in a similar steady state flow, although more calculation time was needed.
After the feeding system is completely filled, the fluid flow reaches steady state after about 0.12 s. Figure 2.10 shows the filling of the flow distributors by means of contour plots of the z-velocity and the velocity difference between the contours is constant. Figure 2.10a shows the filling for feeding system 4 (7/6/1.5mm). The initial state t₁ is a representative moment when the fluid starts to enter the exit slit, the intermediate state t₂ is defined as a moment halfway steady state and t₃ is steady state at 0.12 s. The highest velocities are found at t₁ when the fluid enters the contraction to the holes and to the exit slit. The holes are filled smoothly and the exit slit is filled from the centre (t₁). The initial state also shows that the fluid comes off the walls which is the influence of the surface tension. The intermediate state of the completely filled system shows the influence of the holes. Inside the holes different velocities are present. In the steady state, the velocity profiles are still showing some irregularities inside the distribution chamber and exit slit.

Figure 2.10b shows the filling of the flow distributor for system 5 (7/6/1.5mm). The initial state shows similar results as found for system 4 and the intermediate state shows a few irregularities inside the distribution chamber and exit slit. The steady state situation of the filling shows almost a smooth distribution of the fluid flow although there are still irregularities present. System 6 (7/6/0.5mm) shows a similar filling behaviour as system 5 although solidification of the liquid hinders further flowing from the exit slit. When this occurs, the fluid is recirculating inside the distribution chamber. This corresponds with experiments that were performed with an exit slit thickness of 0.5 mm and it explains why the production of strip stopped.
Figure 2.10. Contour plot of the z-velocity (z-x plane; y=0) of the filling of the flow distributor for system 4 (a) and system 5 (b). $t_1$ is the initial state, $t_2$ is an intermediate state and $t_3$ is steady state. The arrows indicate the positions of the holes. The velocity difference between the contours is constant. Maximum values (m/s) are given in the figure.

All systems show that the fluid leaves the system first from the sidewalls and from the middle of the exit slit. Sometimes this is also experimentally found. The calculations were performed taking into account surface tension. There are no differences found when the simulations were performed without taking into account surface tension.
2.5 Discussion

1. Velocity differences between the feeding systems
The internal velocities of systems 4 (7/Ø6/1.5mm) and 5 (⌀/Ø6/1.5mm) simulated by Flow3D® differ from each other due to the different flow area of the distribution chamber. A relatively decrease of area should correspond to a relatively increase of velocity for each system. Since a constant flow rate exists (0.41x 10^-4 m³/s), a straight line is expected for the average z-velocity in the centre of the systems calculated with an average exit velocity of 0.6 m/s (Q=v₁A₁=v₂A₂).

![Figure 2.11. Average z-velocity in the centre of area C₁; system 4(□), system 5(△); Area C₂; system 4(▲), system 5(■); The exit slit (x). Expected from a constant flow rate of 0.41x10^-4 m³/s(—).](image)

In Figure 2.11 the expected velocity is compared to the average z-velocity in the centre of area C₁, C₂, and the exit slit as obtained from simulations. As can be seen, the transition to C₁ gives for both systems the expected average velocity. At the end of the holes (C₂) the velocity increases due to the development of a pronounced velocity boundary layer. It is seen that system 5, without flow modifier, has the smallest velocity difference between the two C positions. This is attributed by the vortexes (see Figure 2.8) that hinder the velocity in the z direction.

2. Friction losses inside the systems
Inside the feeding system there is a gradual contraction towards the distribution chamber and a sharp contraction towards the exit slit. After the exit slit the fluid
experiences an expansion. Table 3 shows the friction losses in the flow distributor that are calculated following the procedure in [15].

<table>
<thead>
<tr>
<th></th>
<th>$E_f$ walls ($m^2/s^2$)</th>
<th>$E_f$ contractions ($m^2/s^2$)</th>
<th>$E_f$ expansion ($m^2/s^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>System 4</td>
<td>0.02</td>
<td>0.086</td>
<td>0.15</td>
</tr>
<tr>
<td>System 5</td>
<td>0.01</td>
<td>0.084</td>
<td>0.15</td>
</tr>
</tbody>
</table>

Table 2.4 shows that the highest friction losses are caused by the expansion of the fluid. For both systems they are equal because the exit velocities obtained by Flow3D® are equal and the geometric ratios at the outlet area are equal. According to Table 2.4 the differences in friction losses between the systems are negligible. This means that the flow modifier has no effect on the friction losses and therefore on the flow rate.

3. Effect of flow modifier

The flow modifier results in different filling behaviour of the systems which corresponds with literature [6, 8]. The fluid fills first the holes of the flow modifier and then it feeds the exit slit. The residence time for both systems is determined by tracing markers on different positions. The systems give marginal differences and the residence times are $0.33 \pm 0.02$ s and $0.39 \pm 0.02$ s for system 4 (7/Ø6/1.5mm) and system 5 (α/Ø6/1.5mm) respectively. In contrast to [6] the two feeding systems do not give rise to larger residence times in the steady state condition.

The velocity profiles of the three simulated systems are almost smooth. In contrast to [11, 12] our feeding system without flow modifier does not result in a smooth flow in practice, since solidification also plays an important role. Thus a smooth velocity profile is not the only criterion for optimising feeding systems.

From simulations it is seen that the flow modifier introduces velocity in the x-directions and the fluid recirculates inside each hole. This leads to two possible effects. The first effect is that the recirculation in each hole guarantees that the liquid remains evenly distributed in the distribution chamber. This gives a constant feeding of the exit slit. Without flow modifier local recirculations are present leading to an uncontrolled flow in the distribution chamber. As a consequence, the exit slit is not filled with a constant velocity and this will lead to local differences in the strip thickness. These spots have a weak coalescence due to solidification shrinkage and this can result in gaps. A second effect is that circulations in each hole of the flow modifier prevents dead zones. Dead zones cause agglomeration of inclusions [8].
Recirculation of the liquid inside the holes might therefore desirable and exerts a negative influence on the agglomeration of oxides and inclusions.

In case of the experiments the effect of the flow modifier was not demonstrated. The feeding systems with an exit slit of 0.5 mm gave no strip and from simulations with Flow3D® it is confirmed that it is caused by fast heat extraction from the wheel. The one with an exit slit thickness of 1.5 mm gave strip without gaps and is applied with a flow modifier. From simulations it is concluded that the flow modifier gives a constant feeding of the exit slit.

2.6 Conclusion

- From experiments it is concluded that strip of good quality is produced with feeding system 4 consisting of a flow modifier with 7 holes (Ø 6 mm) and an exit slit of 1.5 mm.
- Simulations could not confirm the differences in the flow from the exit slit between feeding systems.
- Simulations show that inside the feeding system the flow modifier introduces recirculations of the fluid which leads to a constant feeding of the exit slit.

References

Strip Casting of Aluminium Alloys

3 Formation of thin strip

3.1 Introduction

Strip casting [1-4] is a continuous casting process designed to produce thin strip or sheet material with gauges of 1 to 25 mm. It is therefore a near net shape casting technique and the cooling kinetics resembles that of related casting processes (twin-roll casting, planar flow casting, etc.). Strip casting results in high cooling rates, which offers substantial metallurgical advantages including an increased solid solution level, a refined microstructure, and consequently the possibility to create dedicated alloys. The strip is formed when liquid metal contacts a rotating wheel by forcing the liquid through an exit opening of a delivery system (Figure 3.1).

![Figure 3.1. Strip formation of liquid metal contacting a rotating wheel (constraint flow).](image)

The liquid contacts the wheel and the melt flows between the wheel and the bottom of the delivery system in a constraint manner [5-10] or in a free jet manner [11-13]. In constraint flow the liquid contacts both the bottom of the feeding system and the wheel (see Figure 3.1) whereas in free jet flow the fluid only contacts the wheel. The type of flow condition is determined by wheel velocity, velocity of the liquid, gap distance and exit slit thickness. In single-roll strip casting it is not directly known which condition will occur. The mentioned parameters influence the thickness of the strip beside process conditions such as casting temperature, wheel properties, hydrostatic pressure, etc. There are three main approaches to derive the
Strip Casting of Aluminium Alloys

thickness of the final strip [4-9, 11, 14] and it is found that strip thickness varies inversely proportional with velocity.

First approach is that the thickness \((d)\) is momentum transport controlled, in which the thickness is determined by the velocity at the exit slit of the feeding system. From the Bernoulli equation [14] and taking into account a constant mass flow, the thickness of the strip can be calculated with the following equation:

\[
d = C \frac{d_{slit} \sqrt{2g_z \Delta z}}{v_w} \quad (3.1)
\]

with \(g_z \Delta z\) is the potential energy, \(C\) is a constant representing friction losses, \(v_w\) is the velocity of the rotating wheel which is equal to the velocity of the strip assuming no slip between wheel and strip, \(d_{slit}\) is the exit slit thickness. Mostly the friction losses are related to the ratio between gap distance \((g)\) and exit slit thickness \((d_{slit})\) but values vary much for the different casting processes and casting conditions. Under free jet flow conditions, it is experimentally found that the thickness will be mainly controlled by the exit slit thickness if \(g \geq d_{slit}\) [7, 8]. Then \(C\) becomes 2/3 which is the friction loss of the fluid when it passes the rectangular exit slit. When the gap distance is smaller than the exit slit thickness, \(C\) becomes [7, 8]:

\[
C = \frac{2}{3} \left( \frac{g}{d_{slit}} \right)^{0.25} \quad (3.2)
\]

Second approach is that the thickness is heat transport controlled, in which the strip thickness is entirely determined by the contact time \((t)\) due to solidification [14]:

\[
d = \frac{h(T_M - T_0)}{\rho' H_f a} t - \frac{h}{2k'} d^2 \quad (3.3)
\]

where \(h\) is the heat transfer coefficient, \(Tm\) and \(To\) are the melting and substrate temperature respectively, \(H_f\) is the latent heat, \(\rho'\) is the density, and \(k'\) is the conductivity and \(a\) is a factor determined by:

\[
a = \frac{1}{2} + \sqrt{\frac{1}{4} + \frac{C_p (T_M - T_0)}{3H_f}} \quad (3.4)
\]
where $C'p$ is the specific heat and primes indicate values of the solid metal. The last part of Equation 3.3 can be ignored if the thickness is small.

Third approach is that the thickness of the strip is controlled by a combination of fluid flow and heat transport which is possible due to the availability of numerical methods [15-18]. This approach calculates a boundary layer thickness as a consequence of fluid flow and thermal properties.

Since the strip thickness is influenced differently for each casting process, in this study the single-roll strip cast process is simulated. By varying process properties, it is investigated which flow condition will occur and whether it influences the strip thickness. Also the surface quality is influenced by these process parameters [10, 12, 13]. Research in this area is mostly limited to analysis of the flow patterns on both surfaces [9] or analysis of instabilities at both menisci which is determined by surface turbulence. The dimensionless Weber number ($We$) gives the conditions for surface turbulence:

$$We = \frac{v_w^2 \rho g}{2\sigma} \quad (3.5)$$

where $\sigma$ is the surface tension, $g$ is the gap distance, $v_w$ is the velocity of the wheel, and $\rho$ is the density of the liquid. This number is the ratio of two pressures which are acting on the surface of the liquid: the inertial pressure and the pressure due to surface tension. When the inertial forces are larger than the surface tension forces the liquid flow falls apart in droplets. The limits of this number to define turbulence are not exactly known although [19] suggests that $We$ numbers in the range of 0.2-0.8 define the maximum value for flow conditions that are free from surface turbulence.

In this study the formation of the strip is simulated with Flow3D® to determine the influence of process parameters on the thickness of the strip. Also some experiments on a single-roll strip caster are performed. The simulations and experiments are also used to study the quality of the surfaces.
3.2 Numerical model

The material flow from the exit slit of the feeding system onto the wheel of our single-roll strip caster is simulated with a commercially available CFD program *FLOW-3D®*. The simulation is based on 2-dimensional transport with laminar fluid flow from the exit slit of the feeding system. Laminar flow is assumed since in practice only a laminar flow results in thin strip with a constant thickness. The columnar height of the liquid inside the feeding system provides the fluid flow from the exit slit. In the current simulation the specific geometry of the feeding system is eliminated by taking into account a pressure at the exit slit. It is assumed that the curvature of the roll is negligible because the thickness of the strip, approximately 1-3 mm, is much smaller than the radius of the wheel (0.15 m) and therefore the substrate is supposed to be flat. Solidification starts when the liquid contacts the wheel, then the temperature in the grid cell drops. The solid material moves with the same velocity as the wheel. The geometry of the simulation is schematically presented in Figure 3.2. The simulated substrate length is taken at least 80 mm and exists than of 100 cells. The cell length and cell height is always taken with a ratio of 1 to 3 which results in a gradual local refinement of the cells in the exit slit [20]. The gap distance is taken 1-3 mm and is much smaller than the substrate length. The vertical direction exists of at least 30 grid cells.

![Figure 3.2. Schematically presentation of the geometry. P: pressure; Tc: casting temperature; g: gap distance; del: exit slit thickness; ve: linear wheel velocity; h: heat transfer coefficient.](image)

The fluid is incompressible and there are no heat losses from liquid towards feeding system or surrounding air. The heat transfer coefficient between liquid and wheel is taken constant and is based on experiments as described in Appendix A. Shrinkage during solidification is not taken into account in this simulation and the surface tension is taken constant. The viscosity depends on temperature which means that the material has a viscosity of 1 when it is solid. The process parameters
and the material properties representative for the AlMn1Mg1 alloy are given in Table 3.1. The parameters that are imposed are pressure (columnar height), gap distance, wheel velocity, temperature of the melt, heat transfer coefficient, and the thickness of the exit slit. The range of variation of the different parameters is listed in Table 3.2. From the simulation the strip thickness and the length of the solidification zone (Figure 3.1) are determined. The solidification zone is the length in which a solid layer builds up a certain strip thickness.

Table 3.1. Process parameters and material properties.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value 1</th>
<th>Value 2</th>
</tr>
</thead>
<tbody>
<tr>
<td>contact angle</td>
<td>140</td>
<td>degree</td>
</tr>
<tr>
<td>latent heat</td>
<td>390</td>
<td>kJ kg(^{-1})</td>
</tr>
<tr>
<td>liquidus temperature</td>
<td>932.5</td>
<td>K</td>
</tr>
<tr>
<td>solidus temperature</td>
<td>932</td>
<td>K</td>
</tr>
<tr>
<td>viscosity at melting point</td>
<td>0.0013</td>
<td>kg m(^{-1}) s(^{-1})</td>
</tr>
<tr>
<td>surface tension</td>
<td>0.84</td>
<td>N m(^{-1})</td>
</tr>
<tr>
<td>specific heat</td>
<td>900</td>
<td>J kg(^{-1}) K(^{-1}) (solid)</td>
</tr>
<tr>
<td></td>
<td>1090</td>
<td>J kg(^{-1}) K(^{-1}) (liquid)</td>
</tr>
<tr>
<td>thermal conductivity</td>
<td>156</td>
<td>W m(^{-1}) K(^{-1}) (solid)</td>
</tr>
<tr>
<td></td>
<td>98</td>
<td>W m(^{-1}) K(^{-1}) (liquid)</td>
</tr>
<tr>
<td>density</td>
<td>2700</td>
<td>kg m(^{3}) (solid)</td>
</tr>
<tr>
<td></td>
<td>2368</td>
<td>kg m(^{3}) (liquid)</td>
</tr>
</tbody>
</table>

Table 3.2. The range of parameters and standard used values.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Range</th>
<th>Standard</th>
</tr>
</thead>
<tbody>
<tr>
<td>casting temperature</td>
<td>943-993</td>
<td>973</td>
</tr>
<tr>
<td>pressure</td>
<td>400-2300</td>
<td>850</td>
</tr>
<tr>
<td>exit slit thickness</td>
<td>0.5-3</td>
<td>1.5</td>
</tr>
<tr>
<td>gap distance</td>
<td>1-3</td>
<td>1.5</td>
</tr>
<tr>
<td>linear wheel velocity</td>
<td>0.43-2.5</td>
<td>0.63</td>
</tr>
<tr>
<td>heat transfer coefficient</td>
<td>10-35</td>
<td>35</td>
</tr>
</tbody>
</table>
3.3 Experimental validation

Several thin strip cast experiments were performed with a laboratory single-roll strip caster. Liquid AlMn1Mg1 alloy is poured into a vertical delivery system with an rectangular exit slit. The system is positioned on top of the rotating wheel (CuCr) where five experiments are carried out with a gap distance of 2 mm and six experiments are carried out with a gap distance of 1 mm. For all experiments the exit slit thickness is 1.5 mm, the columnar height of the liquid is 100 mm, and the casting temperature is 973 K. The cooling strip is moving with a linear velocity that equals the velocity of the wheel, which is 0.63 m/s or 0.8 m/s (Table 3.3).

<table>
<thead>
<tr>
<th>number of experiments</th>
<th>gap distance (mm)</th>
<th>wheel velocity (m/s)</th>
<th>temperature (K)</th>
<th>liquid height (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>5</td>
<td>2</td>
<td>0.63</td>
<td>973</td>
<td>100</td>
</tr>
<tr>
<td>5</td>
<td>1</td>
<td>0.63</td>
<td>973</td>
<td>100</td>
</tr>
<tr>
<td>1</td>
<td>1</td>
<td>0.8</td>
<td>973</td>
<td>100</td>
</tr>
</tbody>
</table>

A flat representative piece of strip from each casting is examined to determine the average thickness. The thickness is determined in two ways; first thickness is measured with a micrometer at several locations and the mean thickness and standard deviation are determined. Second, thickness is calculated from mass \(m\), area \(A\) and density \(\rho\):

\[
d = \frac{m}{\rho A} \quad (3.6)
\]

The uniformity of the thickness is given by the deviation in strip thickness along the strip length. The quality of the strip is determined by visible inspection of flow patterns and by surface roughness measurements. The surface roughness \(R_a\) is measured using a Taylor–Hobson Talystep surface profiler. The profile is measured over a sampling length of 0.8 mm and any low frequency component is filtered out which results in a height contour varying around a zero baseline. The results are compared with the \(We\) number.
3.4 Numerical results

3.4.1 Constraint or free jet flow

The liquid on the wheel flows in a constraint or free jet manner depending on the combination of process parameters. Figure 3.3 shows typical examples of these two flow conditions. In the next section is mentioned which condition occurs for each set of parameters.

![Figure 3.3. Typical examples of free jet flow (a) and constraint flow (b). Black: liquid; grey: solid. In between both states is the solidification zone. The length scale in horizontal and vertical direction is not similar. a) 400 Pa; b) 1700 Pa.](image)

Figure 3.4 shows the velocity profiles inside a puddle for free jet flow conditions and for constraint flow conditions. It is seen that the velocity at the downstream meniscus (see Figure 3.1) is very large in the free jet flow condition. This is not seen for constraint flow and might be a reason for an improved quality of the free surface.
Figure 3.4. Vector plots of the puddle in free jet flow (a) and constraint flow (b). Gap distance: 1.5 mm; pressure: 850 Pa (a), 1700 Pa (b). The length scale in horizontal and vertical direction is not similar.

3.4.2 Effect of process conditions on simulated flow

Figure 3.5 shows the effect of pressure on the thickness of the strip and the length of the solidification zone. The thickness of the strip varies from 0.9 mm with a pressure of 400 Pa to 1.4 mm with a pressure of 2300 Pa. The length of the solidification zone increases in the same way as the strip thickness.
Figure 3.5. Strip thickness (——) and length of solidification zone (----) as a function of pressure. Constraint flow: ⋄; Free jet flow: □.

Figure 3.6. Fluid flow as a function of pressure. a) 400 Pa; b) 850 Pa; c) 1700 Pa; d) 2250 Pa. Gap distance: 1.5 mm; exit slit thickness: 1.5 mm.

Figure 3.6 shows some typical examples of the flow as a function of pressure and it is found that the melt flows in a free jet manner for pressures smaller than or equal to 1150 Pa. When the pressure increases, the volume of the upstream meniscus becomes larger and at a pressure of 1700 Pa which corresponds to a columnar height of 0.073 m, the fluid flows in a constraint way. At higher pressures, 2250 Pa, it is seen that the volume of the upstream meniscus still increases and results in a larger solidified layer directly underneath the exit slit. As a consequence the fluid flow
from the exit slit causes partly remelting of this layer which results in a varying thickness and unstable flow because sometimes the solidified layer becomes as large as the gap distance and that will hinder the fluid flow in practice.

Figure 3.7 shows the strip thickness as a function of linear wheel velocity. As was expected the strip thickness decreases with increasing velocity. The length of the solidification zone remains more or less constant in free flow condition and decreases under constraint flow conditions. In this condition the volume of the downstream meniscus increases in the same way as was seen when the pressure increases. With a low wheel velocity of 0.33 m/s, which gives a constraint flow, the resistance between the moving solidified strip and the feeding system becomes large and in practice, the casting process will be stopped. Typical flow conditions are shown in Figure 3.8.

![Figure 3.7. Strip thickness (---) and length of solidification zone (----) as a function of linear wheel velocity. Constraint flow: ●; Free jet flow: □.](image_url)

![Figure 3.8. Fluid flow as a function of wheel velocity. a) 0.43 m/s; b) 1.5 m/s. Gap distance: 1.5 mm; exit slit thickness: 1.5 mm.](image_url)
Figure 3.9 shows the strip thickness as a function of the casting temperature. It is seen that the thickness of the strip is almost similar for the different temperatures. The length of the solidification zone slightly increases with increasing temperature. Until temperatures of 973 K (700 °C), the melt flows under free jet flow conditions as seen in Figure 3.10a and at temperatures larger than 983 K, it flows constraint such as seen in Figure 3.10b. Apparently, the casting temperature does not effect the thickness or the solidification length although it does effect the fluid flow condition.

![Graph showing strip thickness and length of solidification zone as a function of casting temperature.]

*Figure 3.9. Strip thickness (●) and length of solidification zone (---) as a function of casting temperature. Constraint flow: ●; Free jet flow: □.*

![Diagram showing fluid flow as a function of casting temperature.]

*Figure 3.10. Fluid flow as a function of casting temperature. a) 943 K; b) 993 K. Gap distance: 1.5 mm; exit slit thickness: 1.5 mm.*

Figure 3.11 shows the strip thickness as a function of heat transfer coefficient. It is seen that the thickness of the strip remains constant with increasing heat transfer coefficient. Normally, it is expected that this results in a larger thickness as can be seen from Equation 3.2. This will be discussed in Section 3.6.1. From Figure 3.11 it is seen that the heat transfer coefficient in the constraint flow
condition has a larger effect on the solidification length than in the free jet flow condition. Figure 3.12 shows some typical fluid flows.

Figure 3.11. Strip thickness (----) and length of solidification zone (-----) as a function of heat transfer coefficient. Constraint flow: ●; Free jet flow: □.

Figure 3.12. Fluid flow as a function of heat transfer coefficient. a) 10 kWm$^{-2}$K$^{-1}$; b) 50 kWm$^{-2}$K$^{-1}$. Gap distance: 1.5 mm; exit slit thickness: 1.5 mm.

3.4.3 Effect of geometry on simulated flow

Figure 3.13 shows how strip thickness and solidification length are influenced by the gap distance. It is seen that the strip thickness varies from 0.8 mm to 1.4 mm and that it reaches a maximum with a gap distance of 2 mm. Hereafter the strip thickness decreases again because part of the liquid flows backwards. Similar results are obtained for the length of the solidification zone.
Figure 3.13. Strip thickness (---) and length of solidification zone (-----) as a function of gap distance. Constraint flow: ●; Free jet flow: □.

Figure 3.14. Fluid flow as a function of the gap distance. a) 1 mm; b) 2 mm. Linear wheel velocity of 0.63 m/s; exit slit thickness: 1.5 mm.

Figure 3.14 shows the typical fluid flow behaviour as a function of the gap distance. It is found that constraint flow (Figure 3.14a) occurs for gap distances smaller than 1.5 mm and free jet flow occurs for gap distances equal to or larger than 1.5 mm. It is found that in the free jet flow condition the strip thickness still increases because the volume of the upstream meniscus increases. Gap distances larger than 2 mm result in a flow backwards.
Figure 3.15. Strip thickness (---) and length of solidification zone (----) as a function of exit slit thickness. Constraint flow: ●; Free jet flow: □.

Figure 3.15 shows the effect on strip thickness and solidification length when the exit slit thickness is varied. It is seen that when the exit slit thickness becomes larger the strip thickness decreases. From Equation 3.1 it was expected that the thickness should increase with increasing slit thickness. This will be discussed in Section 3.6.1. Similar observation is valid for the length of the solidification zone. It is found that constraint flow occurs with an exit slit thickness > 1.5 mm. Figure 3.16 shows two typical fluid flow conditions.

Figure 3.16. Fluid flow as a function of the exit slit thickness. a) 1 mm; b) 2.5 mm. Linear wheel velocity of 0.63 m/s; gap distance: 1.5 mm.
3.5 Experimental results

3.5.1 Experimental strip thickness
Several thin strip cast experiments were performed. The six experiments performed with a gap distance of 1 mm resulted in strip with a sufficient surface quality. The five experiments performed with a gap distance of 2 mm resulted in only two useful strips because of severe backflow. Table 3.4 shows the average strip thickness obtained from measurements with the micrometer and from weight calculations (Equation 3.6). The average measured thickness is 1.6 mm which is somewhat larger than the calculated thickness. Different gap distances do not influence the thickness of the strip.

<table>
<thead>
<tr>
<th>gap distance (mm)</th>
<th>Wheel velocity (m/s)</th>
<th>Thickness (mm)</th>
<th>micrometer</th>
<th>calculated (Eq 3.6)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.8</td>
<td>0.7 ± 0.1</td>
<td>0.5 ± 0.01</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>0.63</td>
<td>1.6 ± 0.2</td>
<td>1.3 ± 0.1</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>0.63</td>
<td>1.6 ± 0.2</td>
<td>1.3 ± 0.2</td>
<td></td>
</tr>
</tbody>
</table>

Increasing the linear wheel velocity to 0.8 m/s, results in a calculated strip thickness of 0.5 mm. The smaller strip thickness due to a higher wheel velocity corresponds with the trend found in the simulations although the effect is stronger for the experiments.

3.5.2 Strip quality
Figure 3.17 shows the free surfaces of the strips that result from a gap distance of 1 mm and 2 mm, cast at a velocity of 0.63 m/s. From this figure it is seen that the free surface of strip cast with a gap distance of 1 mm is smoother. Strip cast at a wheel velocity of 0.8 m/s results in a rough surface which looks like grinding paper.

![a) Free surface of strip that is cast with a gap distance of 1 mm](image1)

![b) Free surface of strip that is cast with a gap distance of 2 mm](image2)

*Figure 3.17. Free surface of strip that is cast with a gap distance of 1 mm (a) and 2 mm (b).*
Visual analysis of the strip surface at the wheel side shows that all surfaces cast with a gap distance of 1 mm and a wheel velocity of 0.63 m/s are smooth. The strip that is cast with a wheel velocity of 0.8 m/s shows holes that seem to occur due to air bubbles. Apparently, due to the high wheel velocity air is entrapped in the liquid when it contacts the wheel. This exhibits as a rough surface at the free side and as holes at the wheel side. All surfaces cast with a gap distance of 2 mm show flow lines as can be seen in Figure 3.18. The lines appear white and consist of oxide films that occur due to bad coalescence of the melt. Apparently with a gap distance of 2 mm the liquid is insufficiently compressed which enables the formation of oxide films before the liquid contacts the wheel. These films grow together and give a flow pattern.

Roughness measurements and $We$ numbers (Equation 3.5) are shown in Table 3.5. The surfaces at the wheel side give the smallest roughness compared to the free surfaces. The smoothest wheel surface, found with a gap distance of 1 mm and a wheel velocity of 0.63 m/s, corresponds to a low $We$ number. This is probably due to constraint flow [7].

**Table 3.5. Surface roughness and $We$ numbers of the strips.**

<table>
<thead>
<tr>
<th>gap distance (mm)</th>
<th>wheel velocity (m/s)</th>
<th>roughness $R_a$ (µm)</th>
<th>$We$ number</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>wheel surface</td>
<td>free surface</td>
</tr>
<tr>
<td>1</td>
<td>0.8</td>
<td>2.2±0.4</td>
<td>25*</td>
</tr>
<tr>
<td>1</td>
<td>0.63</td>
<td>1.6±0.6</td>
<td>3.8±1.0</td>
</tr>
<tr>
<td>2</td>
<td>0.63</td>
<td>2.5±1.5</td>
<td>3.2*</td>
</tr>
</tbody>
</table>

* based on 1 measurement.
3.6 Discussion

From the point of view that the numerical model gives the most complete solution of
the influence of different parameters onto the strip thickness, the next sections
discusses the analytical models to see which one comes closest to the simulated
results. In the last section the experiments will be discussed.

3.6.1 Thickness of the strip.

Figure 3.19 shows the momentum transport controlled thickness, calculated from
Equations 3.1 and 3.2, against the thickness obtained from simulations. The
calculated thickness is shown for all simulations with a gap distance and exit slit
thickness of 1.5 mm. It is seen from this figure that the calculated thickness
corresponds well for a simulated thickness of 1 mm and smaller which corresponds
to free jet flow conditions. For a simulated thickness larger than 1 mm the analytical
model mostly overestimates the thickness for free jet flow conditions and for
constraint flow conditions.

![Graph showing thickness comparison](image)

*Figure 3.19. Thickness as calculated from Equations 3.1 and 3.2, against the thickness obtained from simulations for gap distance and exit slit thickness of 1.5 mm. Constraint flow: ●; Free jet flow: □.*

The explanation follows from the fact that Equations 3.1 and 3.2 are not taking into
account solidification. The solid layer results in friction with the liquid flow. It
seems that the contribution of this friction is negligible with a thickness smaller than
1 mm and becomes more pronounced at a thickness larger than 1 mm. With a larger
thickness, the friction between fluid flow and growing solidification layer becomes
larger and than heat transport and contact time dominates the final strip thickness. In this case Equations 3.1 and 3.2 are not longer representative since the contact time can only be influenced with different heat transport coefficients or wheel velocities.

Figure 3.20. Thickness as calculated from Equation 3.1 against the thickness obtained from simulations for an exit slit thickness of 1.5 with varying gap distances. Constraint flow: ●; Free jet flow: □.

Figure 3.21. Thickness as calculated from Equation 3.1 against the thickness obtained from simulations for a gap distance of 1.5 mm with varying exit slit thickness. Constraint flow: ●; Free jet flow: □.
Figure 3.20 and Figure 3.21 show the calculated and simulated thickness as a function of geometry. It is seen for both figures that the thickness calculated from Equation 3.1 reasonably corresponds to the thickness obtained from simulations for free jet flow conditions but does not correspond for constraint flow conditions. An overestimation of the calculated thickness for constraint flow as a result of a larger exit slit thickness (Figure 3.21) follows from the fact that the flow rate increases. With increasing flow rate also the velocity of the liquid onto the wheel increases. Since constraint flow occurs and because the solid layer moves with a constant velocity, the friction between the liquid flow and the solidification layer becomes larger with increasing exit slit thickness resulting in a smaller thickness instead of a larger thickness.

The overestimation of the thickness for constraint flow as a result of smaller gap distances (Figure 3.20) is probably attributed to the friction between wheel and feeding system. Apparently the friction factor should be taken larger. For gap/exit slit ratios < 1 Equation 3.2 becomes:

$$C = \frac{2}{3} \left( \frac{g}{d_{exit}} \right)^{1.37}$$  \hspace{1cm} (3.7)

Equation 3.1 corrected with this factor is represented in Figure 3.22.

![Thickness diagram](image)

*Figure 3.22. Thickness as calculated from Equations 3.1 and 3.7 against the thickness obtained from simulations for different gap distances. Constraint flow: ⬤; Free jet flow: □.*

Figure 3.23 shows the calculated thickness against the simulated thickness when we assume that the thickness is controlled by heat transport. It is seen from
Figure 3.23 that excellent agreement exists between calculated thickness and simulated thickness, for all conditions and geometries under consideration. It also appears that it is in agreement for both flow conditions. However, for the calculated thickness the value for $t$ (Equation 3.3) was derived from the solidification length obtained from the simulations. When the length of the solidification zone is taken constant a broader scattering around the 45°-line is obtained and agreement is limited.

![Graph](image)

*Figure 3.23. Thickness as calculated from Equation 3.3 against the thickness obtained from simulations.*

3.6.2 **Length of the solidification zone and contact time**

Figure 3.24 shows the length of the solidification zone as a function of the thickness obtained from simulations. It shows a broad scattering of data points which indicates that the length of the solidification zone only does not determine the thickness.
Figure 3.24. Length of solidification zone against the simulated thickness

Figure 3.25 shows the contact time (the length of the solidification zone divided by the wheel velocity) against the simulated thickness. A linear relationship is obtained between contact time and thickness for a certain $h$-value. It is seen that good agreement exists between the predictions of Equation 3.3 and the simulations. Heat transfer coefficient hardly effects the thickness which is in agreement with [17].

Figure 3.25. Contact time against simulated thickness. Solid lines give predictions of Equation 3.3. ◇: heat transfer coefficient.

3.6.3 Constraint flow and free jet flow

From the former sections it is clear that constraint flow or free jet flow is not only determined by the gap/exit slit ratio. It is likely that free jet flow occurs when the
gap distance is larger than the exit slit thickness but the flow becomes constraint when a larger liquid flow is applied (or relatively larger liquid flow), and when the heat transport becomes slower. A relatively larger liquid flow results from higher pressures, lower wheel velocities, and higher casting temperatures (lower viscosity). A slower heat transport, that means a smaller heat transport coefficient, results also into a relatively larger liquid flow and thus gives constraint flow.

3.6.4 Validation
With the former sections it is now possible to predict the flow condition in the experimental set up and the strip thickness. It is expected that with a wheel velocity of 0.63m/s, and a gap/exit slit ratio smaller than 1, constraint flow occurs. The experimentally obtained thickness of 1.6 mm comes close to the simulated thickness of 1.41 mm for large pressures. At larger wheel velocities, the experimentally thickness is 0.7 mm which is in agreement with our simulations.

3.7 Conclusions

From simulations is concluded that:

- the strip thickness is determined by the contact time. The heat transport equation predicts the thickness when the contact time between wheel and liquid is known or when the solidification length is known.
- the overall transport equation (Bernoulli) overestimates the thickness, which can be overcome by adjusting the friction factor.
- constraint flow leads to more friction resulting into a smaller thickness than expected.

The surface quality is influenced by the flow condition. Experiments show an improved quality under constraint flow conditions and this is confirmed by simulations which do not show turbulence at the menisci of the puddle.

References


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Numerical Modeling and Simulation’, Proceedings of Materials Solution
20. R. van Tol, Mould Filling of Horizontal Thin-Wall Castings, PhD Thesis,
4 Influence of casting conditions on the as-cast microstructure

4.1 Introduction

The microstructure of strip cast material differs from the conventional cast material due to a higher cooling rate. Since the as-cast microstructure determines for a large part the properties of the final sheet [1-3] it is important to know the structures that occur and how they are influenced by alloying elements and process parameters.

The as-cast structure of rapidly solidified aluminium strip is generally characterised by three zones; the first zone consists of fine crystals at the wheel side, the second zone of columnar crystals, and the third zone of equiaxed crystals. The columnar crystals grow in a constraint way until the solidifying equiaxed crystals form a barrier to further growth. The equiaxed grain size is controlled not only by nucleation but also by remelting of columnar dendrites [4-6].

Some experiments [7, 8] and simulations [9] have been carried out to determine the influence of casting conditions on the length of the columnar zone and the grain size of the equiaxed zone. The experiments and simulations were carried out under planar flow casting conditions for an Al-Cu alloy [7] and an Fe alloy [8, 9]. The major effects are summarised in Table 4.1. Agreement between these authors exist about the influence of casting temperature on the columnar length and about the influence of thickness and heat transfer coefficient on the grain size. Disagreement exists about the influence of heat transfer coefficient on the columnar length and about the influence of the casting temperature on the grain size.

In this research the influence of minor additions of alloying elements, casting temperature, wheel (substrate) temperature, and strip thickness on the as-cast structure of aluminium strip is determined by measuring the length of the columnar zone and the equiaxed grain size, and by examining the morphology of the intermetallics.

To this purpose experiments were performed with the strip cast unit and by pouring the melt on the wheel of the strip cast unit in a static position. The latter experiments were performed to guarantee a constant thickness, since the thickness of the strip during strip casting depends, beside the wheel velocity, on casting temperature and wheel temperature (see Chapter 3).
Table 4.1. Effects of casting conditions on grain size and columnar length. ‘-‘=influence not mentioned.

<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Columnar length</td>
<td>$T_e \uparrow$</td>
<td>-</td>
<td>$T_e \uparrow$</td>
</tr>
<tr>
<td></td>
<td>$d \downarrow$</td>
<td>-</td>
<td>$h \downarrow$</td>
</tr>
<tr>
<td></td>
<td>in combination</td>
<td>-</td>
<td>$T_w \uparrow$</td>
</tr>
<tr>
<td></td>
<td>with $h \uparrow$</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Grain size</td>
<td>$T_e \uparrow$</td>
<td>$T_e \downarrow$</td>
<td>$T_e$ no effect</td>
</tr>
<tr>
<td></td>
<td>$d \uparrow$</td>
<td>-</td>
<td>$d \uparrow$</td>
</tr>
<tr>
<td></td>
<td>in combination</td>
<td>-</td>
<td>$h \downarrow$</td>
</tr>
<tr>
<td></td>
<td>with $h \downarrow$</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td></td>
<td>$A$ $\downarrow$</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

$T_e$: casting temperature, $T_w$: wheel temperature, $h$: heat transfer coefficient, $d$: thickness, $A$: content of alloying elements.

4.2 Experimental

The experiments are carried out with the strip cast unit as described in Chapter 2. The experiments consist of casting on the rotating copper wheel with a feeding system as well as casting on the static copper wheel in a circular mould. The feeding system has a rectangular exit slit with a thickness of 1.5 mm and a gap distance of 1 mm and the circular mould has a diameter of 0.05 m. Further details about the feeding system are described in Chapter 2. Strip casting of aluminium alloys with a superheat of 40 °C and a linear wheel velocity of 0.63 m/s result in a strip with a thickness of 1.4 mm. The wheel substrate has a uniform temperature of 20 °C. The aluminium alloys that are poured on the static wheel maintain a thickness of 10 mm. The melt is poured with a superheat of 0°C or 70°C. The wheel substrate is internally cooled by water and has a uniform controlled temperature of 20°C or 50°C. The aluminium alloys for the experiments have different compositions based on the AA3004 alloy (Table 4.2).

From the strip cast material, samples are taken from the plane normal to the casting direction. This is shown by the hatched area in Figure 4.1. Also samples are taken from the plane parallel to the casting direction and normal to the heat flow, which is shown by the dark grey area in Figure 4.1. From material of the static experiments, cross-sections are taken from the centre of the casting. The samples are
polished to show the intermetallic phases, unless otherwise mentioned, and anodised with 5% HBF$_4$ to show the grains.

<table>
<thead>
<tr>
<th>Composition (wt. %)</th>
<th>Mn</th>
<th>Mg</th>
<th>Si</th>
<th>Fe</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>CP Al**</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMn*</td>
<td>0.8</td>
<td>&lt;0.01</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>0.6</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>1.1</td>
<td>3.9</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg**</td>
<td>1.1</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>1.1</td>
<td>0.8</td>
<td>0.3</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>1.2</td>
<td>0.9</td>
<td>0.1</td>
<td>0.7</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>1.2</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>0.07</td>
</tr>
</tbody>
</table>

* = only used in the static experiments
** = only used in the strip cast experiments

*Figure 4.1. Samples of the strip cast material are taken from the planes normal (hatched area) and parallel (dark grey area) to the casting direction.

4.3 Results

4.3.1 Columnar length and grain size in strip casting
The microstructure of the strip cast alloys show equiaxed grains as can be seen in Figure 4.2. In general, the strip cast material does not contain a zone with columnar crystals in the plane normal to the casting direction (see Figure 4.2).
Strip Casting of Aluminium Alloys

The grain size in the cross-section of the strip depends on alloy composition and ranges from 110 μm to 190 μm, with an average grain size of approximately 150 μm. The CP Al shows an average grain size of 190 μm and the high magnesium containing alloy shows an average grain size of 110 μm. There is not much difference in size between the low manganese containing alloy and the alloys containing silicon, iron or chromium.

Figure 4.2. Complete cross-section of strip cast AlMnMg+Si alloy. Cross-section is normal to the casting direction. Bottom of cross-section: wheel side; top of cross-section: free solidified side.

Figure 4.3. Grains in the plane parallel to the casting direction of strip cast AlMnMg+Cr alloy.

Figure 4.3 shows the grains in the plane parallel to the casting direction of the strip. These grains are generally smaller (approximately 110 μm) than in the plane normal to the casting direction. From this observation it can be concluded that
the crystals are not really equiaxed and that there is some directionality in the direction of the heat flow.

Figure 4.4. Dendritic structure of strip cast AlMnMg+Fe (a) and AlMnMg4 (b). Cross-section is normal to the casting direction. Bottom of cross-section: wheel side. Etched with Keller and Wilcox: Interferential contrast microscopy.

Figure 4.4 shows the dendritic structure which is observed for grains at the wheel side. A directional dendritic structure is found except for the AlMnMg4 alloy where no directionality is observed, as can be seen in Figure 4.4b. The length of these dendrites varies along the wheel side of the strip from a few microns to 0.5 mm.

4.3.2 Columnar length and grain size in static experiments

Figure 4.5a shows the macrostructure of a representative alloy cast at liquidus temperature (superheat: 0 °C) onto the wheel at room temperature. All alloys show a zone with columnar crystals and a zone with equiaxed crystals. A clear transition of columnar crystals to equiaxed crystals is found under these conditions. The length of the columnar crystals is given in Table 4.3. Figure 4.5b and Figure 4.5c show the macrostructures of a representative alloy cast with a superheated melt (70 °C). Since there are equiaxed grains observed between the columnar grains it appears that there is a transition zone rather than a clear boundary. The length of the columnar crystals increases for all alloys when casting with a superheated melt as can be seen in Table 4.3.
Figure 4.5. The columnar zone at the wheel side of the AlMnMg+Cr alloy. Consecutive photographs. (a) Wheel temperature: 20 °C, superheat: 0 °C; (b) Wheel temperature: 20 °C, superheat: 70 °C; (c) Wheel temperature: 50 °C, superheat: 70 °C.

Table 4.3. Columnar length (CL), equiaxed grain size (GS) and Secondary Dendrite Arm Spacing (SDAS) found in the static experiments.

<table>
<thead>
<tr>
<th></th>
<th>Wheel temperature: 20 °C</th>
<th>Wheel temperature: 50 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Superheat: 0°C</td>
<td>Superheat: 70°C</td>
</tr>
<tr>
<td></td>
<td>CL (mm)</td>
<td>GS (µm)</td>
</tr>
<tr>
<td>AlMn</td>
<td>1.5</td>
<td>100</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>1.5</td>
<td>100</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>0.5</td>
<td>130</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>2.0</td>
<td>90</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>1.5</td>
<td>80</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>2.0</td>
<td>100</td>
</tr>
</tbody>
</table>

If not stated SDAS could not be detected with SEM.
Figure 4.6. The equiaxed structures of AlMn0.6Mg as a function of superheat and wheel temperature. a) Superheat: 0 °C; Wheel temperature 20 °C; b) Superheat: 70 °C; Wheel temperature 20 °C; c) Superheat: 70 °C; Wheel temperature 50 °C.

Figure 4.6 shows the typical effect of wheel temperature and superheat on the grain size in the equiaxed zone. The average grain size in the equiaxed zone is about 100 μm when casting with a superheat of 0 °C, which is approximately constant for the various alloys. The grain size becomes much larger for each alloy when casting with a superheat of 70 °C. The largest grain sizes are obtained with the Cr containing alloy which is confirmed by [3] (see Table 4.3).

The morphologies of the interfaces of the equiaxed grains are cellular when casting with a superheat of 0 °C, such as seen in Figure 4.6a, whereas the equiaxed grains are dendritic when casting with a superheat of 70 °C such as seen in Figure 4.6b and c. The spacing between the secondary dendrite arms (SDAS) is only found in material cast with a superheat and is given in Table 4.3. There is practically no difference in SDAS values between material cast on a wheel with a temperature of 20 °C or 50 °C.
4.3.3 The intermetallic phases in material from strip cast experiments

The solidification of AlMnMg alloys involves the formation of several types of intermetallics. Figure 4.7 shows representative examples for all morphologies found which are given in Table 4.4.

Figure 4.7. Morphologies of the Al₁₅(MnFe)₅Si₂ intermetallic. a) AlMnMg₄: plates; b) AlMnMg: long tailed Chinese script; c) AlMnMg+Si: rounded needles. Cross-section normal to the casting direction.

Al₁₆(FeMn) phases are recognised as light grey phases and from Table 4.4 it is seen that the Al₁₆(FeMn) phase, which is the first phase that forms during solidification of an AlMn1Mg1 alloy [3, 10], is generally not observed for the strip cast material. It cannot be excluded that in the AlMnMg+Fe alloy this phase is
formed but due to their small size the phases could not be identified. It is reported that with increased cooling rate these phases become smaller.

Table 4.4. Morphology of intermetallic phases in material from strip cast material.

<table>
<thead>
<tr>
<th>alloy</th>
<th>Al₆(MnFe)</th>
<th>Al₁₅(MnFe)₂Si₂; Al₃(FeMn)</th>
<th>Mg₂Si</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>plate</td>
<td>Chinese script</td>
<td>Plates</td>
</tr>
<tr>
<td></td>
<td></td>
<td>long tailed</td>
<td>short tailed</td>
</tr>
<tr>
<td>CP Al</td>
<td>x</td>
<td></td>
<td></td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>x</td>
<td></td>
<td>x</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td></td>
<td>x</td>
<td></td>
</tr>
<tr>
<td>AlMnMg</td>
<td>x</td>
<td></td>
<td></td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td></td>
<td>x</td>
<td>x</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>x*</td>
<td></td>
<td>x*</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td></td>
<td>x</td>
<td>x</td>
</tr>
</tbody>
</table>

*: Identification of the phases is not reliable due to small size.

The Al₁₅(MnFe)₂Si₂ phase, which can be formed directly from the melt in a AlMn1Mg1 alloy [10] is recognised as a light grey phase with a dark grey edge. Alternatively, this phase may transform from the Al₆(FeMn) phase in case of fast cooling rates. The Al₁₅(MnFe)₂Si₂ phase is observed in all our AlMnMg alloys. The common morphologies of this phase are Chinese script or skeletons, plates, and needles. It is seen that Chinese script is found for alloys with a low content of alloying elements. The curly phase is described in Section 4.3.4. It is reported [3, 10] that the solidification of an AlMnMg4 alloy involves the formation of Al₃(FeMn), Al₃Mg₅, and Mg₂Si phases beside the phases mentioned before. Al₃(FeMn) phases are recognised as dark grey phases. Al₃Mg₅ cannot easily be distinguished from Al₃(FeMn) phases. Plates are the common morphology found, which is confirmed for the Al₆(FeMn) phases [11]. The intermetallic Mg₂Si is found as a dispersion of black particles.

The size of the intermetallics differs between the alloys as can be seen in Figure 4.7. The size of the needles and plates is approximately 1.5 μm and the size of the long tailed Chinese script is approximately 14 μm.

Figure 4.8 shows a representative example of intermetallics as shown in the plane parallel to the casting direction. These cross-sections show in most cases larger phases than those in the plane normal to the casting direction, compare Figure 4.8 with Figure 4.7c. One experiment is carried out with the AlMnMg+Si alloy, cast
on a heated wheel of 50 °C. It is found that distribution and size of phases become equal to those phases from the cross-section of the plane parallel to the casting direction (Figure 4.8).

\[ \text{Figure 4.8. Intermetallics of the AlMnMg+Si alloy. Cross-section parallel to the casting direction.} \]

4.3.4 The intermetallic phases in material from static experiments

The morphologies of the intermetallic phases, present in the material of the static experiments, are listed in Table 4.5. The Al₆(FeMn) phase, such as shown in Figure 4.9, is observed in some alloys and is present as a plate. It is found that the alloy with the lowest concentration of alloying elements and the alloy that contains Fe have coarse Al₆(FeMn) phases. In addition, the AlMnMg+Fe alloy also shows coarse plates of Al₁₅(FeMn)₃Si₂. Apparently large Al₆(FeMn) phases give also large Al₁₅(FeMn)₃Si₂ phases during transformation. Further, the Al₁₅(FeMn)₃Si₂ phase shows the common morphologies, whereas it is observed that the Chinese script morphology (Figure 4.10a) is more developed, compare Figure 4.10a with Figure 4.7b. The alloys with Chinese script are the same alloys found in Table 4.4. Also a curly morphology is found, as shown in Figure 4.10b, which is clearly present in the AlMnMg4 alloy for all conditions.
### Table 4.5. Morphology of intermetallic phases in material from static experiments.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>T&lt;sub&gt;wheel&lt;/sub&gt;</th>
<th>Superheat</th>
<th>Al&lt;sub&gt;4&lt;/sub&gt;(FeMn) Plate</th>
<th>Al&lt;sub&gt;13&lt;/sub&gt;(MnFe&lt;sub&gt;2&lt;/sub&gt;)Si&lt;sub&gt;2&lt;/sub&gt; Plate</th>
<th>MgSi Phase</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>Chinese script Long tailed</td>
<td>Chinese script Short tailed</td>
<td>Rounded needle/particle Curly phase</td>
</tr>
<tr>
<td>AlMn</td>
<td>20°C</td>
<td>0°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td>50°C</td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>20°C</td>
<td>0°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td>50°C</td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>20°C</td>
<td>0°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td>50°C</td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>20°C</td>
<td>0°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td>50°C</td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>20°C</td>
<td>0°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td>50°C</td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>20°C</td>
<td>0°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
<tr>
<td></td>
<td>50°C</td>
<td>70°C</td>
<td>x</td>
<td>x</td>
<td></td>
</tr>
</tbody>
</table>

**Figure 4.9. Morphology of the Al<sub>4</sub>(FeMn) intermetallic (1). Superheat: 70 °C, wheel temperature: 20 °C. AlMnMg+Fe alloy.**
Figure 4.10. Morphology of intermetallics $\text{Al}_3(\text{FeMn})_2\text{Si}_2$ (2) and $\text{Mg}_2\text{Si}$ (3). (a) $\text{AlMnMg+Si}$: Chinese script. Superheat: 0 °C, wheel temperature: 20 °C; (b) $\text{AlMnMg4}$: curly phases. Superheat: 0 °C, wheel temperature: 20 °C.

For almost all alloys it applies that the size of the phases becomes smaller and the distribution of the phases becomes more dispersed when casting with a superheat and increased wheel temperature [3]. The smallest phases are found for the needle morphology which have a size of approximately 2.5 μm and are almost always found with casting with a superheat of 70 °C and a heated wheel of 50 °C. The $\text{Mg}_2\text{Si}$ phases are clustered in groups when casting with superheat of 0 °C and they become more dispersed when casting with a superheat of 70 °C.

4.4 Discussion

4.4.1 Cooling rate
The mean cooling rate of the strip in case of directional solidification can be calculated from the thickness of the strip $d$ [12]:

$$d = \frac{h\Delta T}{\rho H_f a} t - \frac{h}{2k} d^2$$  \hspace{1cm} (4.1)

where $h$ is the heat transfer coefficient taken constant, $\Delta T$ is the temperature difference between melt and wheel, $\rho$ is the density of the solid, $H_f$ is the latent heat of fusion, $t$ is the solidification time, and $a$ is a factor determined by:
Influence of casting conditions on the as-cast microstructure

\[ a = \frac{1}{2} + \sqrt{\frac{1}{4} + \frac{C_p \Delta T}{3H_f}} \]  

(4.2)

After differentiating Equation 4.1, the growth rate \( V \) of the solidification front is obtained and after multiplication of \( V \) with the thermal gradient \( G \), the cooling rate \( R \) becomes:

\[ R = V \times G = \frac{h \Delta T}{\sqrt{1 + 2 \frac{h^2 \Delta T t}{k \rho H_f a \rho H_f a}}} \]  

(4.3)

From this equation it can be seen that, provided that the heat transfer coefficient is constant and independent of the wheel velocity, the mean cooling rate for thinner strip is larger and should approximately differ a factor of 10.

**Table 4.6. Average SDAS of dendrites and local cooling rate.**

<table>
<thead>
<tr>
<th>SDAS (µm)</th>
<th>( R_{\text{local}} ) (K/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 mm</td>
<td>8.5±2</td>
</tr>
<tr>
<td>10 mm</td>
<td>24±6</td>
</tr>
</tbody>
</table>

The local solidification rate can be calculated from measurements on the SDAS [10]:

\[ \log SDAS = 2.04 - 0.38 \log \frac{dT}{dt} \]  

(4.4)

Table 4.6 shows the local solidification rates calculated from the SDAS values in the centre of the strip. The difference of local cooling rates between the two strips is a factor 15. This difference is within experimental errors similar to the factor 10, found with Equation 4.3 and assuming a constant heat transfer coefficient. Therefore, the conclusion that the heat transfer coefficient depends on the wheel velocity is not confirmed [7].

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4.4.2 Columnar length

The differences in columnar length, found between the alloys, are attributed to differences in concentration of elements. This can be demonstrated with the criterion of constitutional undercooling. Constitutional undercooling ahead of the dendrite tip leads to a positive driving force $\Theta$ which causes any perturbation to grow. Such a perturbation hinders further planar growth and causes the formation of secondary dendrites. Consequently the columnar length becomes smaller with increased $\Theta$. The interface of a planar solidification front is constitutionally undercooled when:

$$m_i G_{c,i} > G \text{ and } G > 0 \quad \quad (4.5)$$

$G$ is the temperature gradient, $G_{c,i}$ is the concentration gradient of element $i$ of the binary alloy, and $m_i$ is the liquidus slope [5]. The driving force $\Theta_i$ is:

$$\Theta_i = m_i G_{c,i} - G \quad \quad (4.6)$$

with

$$m_i G_{c,i} = \frac{m_i C_i (k_i - 1) V}{k_i D_i} \quad \quad (4.7)$$

where $C_i$ is the initial alloy concentration, $k_i$ is the distribution coefficient of alloying element $i$ in Al and is determined from the binary phase diagram, and $D_i$ is the diffusion coefficient in the liquid, taken as $10^{-9} \text{ m}^2/\text{s}$ for all alloying elements. The growth rate of the solidification front $V$ is based on the cooling rate as calculated in Section 4.4.1 and taking into account that the strip of 1 mm thickness solidified 10 times faster. For a multicomponent system Equation 4.7 is modified as follows:

$$m G_c = \frac{\sum (m_i C_i (k_i - 1)) V}{<k>D} \quad \quad (4.8)$$

where $<k>$ is a weighted distribution coefficient depending on the alloy composition:

$$<k> = \frac{\sum C_i k_i}{\sum C_i} \quad \quad (4.9)$$

Part of the right hand side of Equation 4.8 is called the growth restriction factor (GRF)[14, 15]:

---

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$$\text{GRF} = \sum m_i C_i (k_i - 1) \quad (4.10)$$

The GRF determines the difference in grain growth due to alloying elements. Table 4.7 shows the values for the liquidus slope and distribution coefficient whereas the alloying elements are listed in decreasing effect on the grain growth as can be shown in the fourth column. Table 4.8 shows calculated values of the growth restriction factor, weighted distribution coefficient, and driving force $\Theta$.

| Table 4.7. Values of the liquidus slope ($m_i$) and distribution coefficient ($k_i$). Data taken from phase diagrams of binary Al alloys. |
|---|---|---|
|   | $k_i$ | $m_i$ (K/wt %) | $(k_i-1)m_i$ (K/wt %) |
| Si | 0.11 | -6.6 | 5.87 |
| Cr | 2.0  | 3.5  | 3.50 |
| Mg | 0.51 | -6.2 | 3.04 |
| Fe | 0.02 | -3   | 2.94 |
| Mn | 0.94 | -1.6 | 0.10 |

Table 4.8. Growth restriction factor, weighted distribution coefficient, and driving force for constitutional undercooling.

<table>
<thead>
<tr>
<th></th>
<th>GRF (K)</th>
<th>$&lt;k&gt;$</th>
<th>$\Theta$ ($\times 10^5$ K/m)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>1 mm</td>
</tr>
<tr>
<td>AlMn</td>
<td>1.25</td>
<td>0.70</td>
<td>2</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>4.57</td>
<td>0.57</td>
<td>31</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>11.61</td>
<td>0.58</td>
<td>89</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>4.89</td>
<td>0.62</td>
<td>31</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>5.49</td>
<td>0.56</td>
<td>40</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>4.88</td>
<td>0.69</td>
<td>27</td>
</tr>
</tbody>
</table>

After applying Equation 4.8 for a multicomponent system it is found that for all alloys $\Theta$ is positive which means that constitutional undercooling is present. Figure 4.11 shows the columnar length found for the 10 mm strip plotted against $\Theta$. An increase of alloy content leads to an increase of driving force and consequently to a smaller columnar length. This trend is indicated by the lines. It is also found that superheating the melt increases the columnar length since it takes more time with a
superheated melt to remove the heat from the melt and to form nuclei for equiaxed grains. In this study, this is confirmed for all alloys. The larger columnar length found with superheating the melt is confirmed by literature [7, 9]. Also it is seen that casting with a superheated melt gives a more pronounced difference in columnar length for low Θ values (*i.e.* the AlMn alloy).

![Graph showing columnar length plotted against constitutional undercooling for 10 mm strip. Superheat 0 °C (○) and 70 °C (▲). Lines indicate the trend.](image)

Table 4.8 shows for the 1 mm strip the driving forces under the assumption that the same ΔT (Equation 4.3) applies as in the 10 mm strip. The driving forces are somewhat lower than for the 10 mm strip. According to the trend found in Figure 4.11, a columnar length is expected. However, the microstructure of the 1 mm strip does not show columnar grains. Figure 4.4 shows that instead of columnar grains directional dendrites are observed, except for the AlMnMg4 alloy where no directionality is observed, as can be seen in Figure 4.4b. These observations correspond with the trend found in Figure 4.11. The irregularities of the length of the dendrites indicate that contact with the wheel varies and is not always optimal. Optimal contact causes nuclei to grow out in the reverse direction of the heat flow. During the competitive growth of the dendrites, the one with optimal contact with the substrate becomes dominant since the constitutional undercooling remains than large. However, this does not explain why the grain is equiaxed instead of columnar. Equiaxed grain formation is promoted by many factors such as temperature fluctuations and alloying content. In our strip cast experiments we assume that forced convection is the main reason for the changing temperature and concentration ahead of the solidification front which promotes grain growth in other directions.
4.4.3 Grain morphology and size of equiaxed grains

Figure 4.12 shows the effect of thermal gradient and growth rate on the morphology and size of the grains. The morphology of the interface of the grains is a function of growth rate \( V \) and temperature gradient \( G \) [5] as can be seen for a positive gradient in Figure 4.12. The ratio \( G/V \) determines the morphology and the product of \( G*V \) (= \( R \)) determines the scale of the grains. The cellular morphology found for the static castings with a superheat of 0 °C corresponds with the indicated position in Figure 4.12. A larger superheat of 70 °C results in a dendritic morphology and an increased cooling rate since secondary dendrites and smaller intermetallics [5] are found. Apparently, casting with a larger superheat has minor influence on the thermal gradient and gives a strongly increased growth rate. The arrow in Figure 4.12 indicates the change in position when casting with a larger superheat.

![Diagram showing grain morphology and growth rate](image)

*Figure 4.12. Solidification morphologies as a function of thermal gradient and growth rate. Data taken from [5]. ● = superheat 0 °C. The arrow (bold) indicates the effect of casting with a superheat.*

It is found that the grain size in the equiaxed zone increases with higher superheat. This is confirmed by actual experiments [7] but contradicts with simulations [9] and other experiments [8]. The combination of a larger grain size and higher cooling rate seems to be contradictive (Figure 4.12) but is explained with the remelting of grains...
caused by the released latent heat of other growing grains [5] which was not accounted for in [9].

![Graph showing grain size as a function of growth restriction factor and superheat.](image)

*Figure 4.13. Grain size as a function of growth restriction factor and superheat. Superheat: 0 °C, wheel temperature: 20 °C (○); superheat: 70 °C, wheel temperature: 20 °C (■); superheat: 70 °C, wheel temperature: 50 °C (△). Lines indicate the trend.*

The grain size at the moment of impingement is determined by the growth restriction factor (GRF) [13-16]. This factor determines the growth restriction for equiaxed grains caused by the solute concentration at the growth front of the dendrite tip. Increasing solute concentration at the growth front will decrease the growth rate of the grain. Large GRF values lead to large restrictions and consequently to small grains. GRF's are given in Table 4.8 and in Figure 4.13 the grain size of material from static experiments is plotted against the growth restriction factor. Figure 4.13 demonstrates that the grain size decreases with increasing growth restriction factor which is in agreement with [15, 16]. Additions of Mg give the smallest grain which is in agreement with [16] who found a decreasing grain size with increasing addition of Mg up to a growth restriction factor of 35 K. However, casting with a superheat of 0 °C shows little effect of the growth restriction factor on the grain size. In this case, where cellular growth was found, the number of nuclei dominates the grain size and the effect of growth restriction by alloying elements is minimal. For strip cast material an equiaxed grain is found which size is in the same range as was found for the 10 mm strip cast with a superheat of 0 °C. It is concluded that the grain size of the strip cast material is determined by remelting of grains.
since the cooling rate is larger than in the 10 mm strip cast with a superheat of 0 °C. Consequently the effect of the GRF is minimal or has disappeared.

4.5 Conclusions

Optical microscopy of various strip cast AlMnMg alloys with a thickness of 1 mm shows that:
- the grains are equiaxed throughout the whole strip,
- the average grain size is approximately 150 μm,
- the size of intermetallic phases ranges from 1 μm to 14 μm.

SDAS measurements show that this strip solidifies with a cooling rate of 800 K/s.

Optical microscopy of various AlMnMg alloys cast with a thickness of 10 mm shows that:
- at the wheel side columnar crystals are present, the lengths of which depend on alloy composition,
- the average grain size ranges from 100 μm (superheat 0 °C) to 600 μm (superheat 70 °C),
- the size of intermetallic phases ranges from 2.5 μm to 50 μm (superheat 0 °C) and from 2.5 μm to 25 μm (superheat 70 °C).

SDAS measurements show that this strip solidifies with a cooling rate of 50 K/s (superheat 70 °C).

Remelting of the growing grains cause the relatively large grain size for the strip cast material (1 mm). The constitutional undercooling determines the differences in columnar length of different alloy compositions. The equiaxed grain size for material with a thickness of 10 mm is determined by the growth restriction factor (superheat 70 °C) or by the number of nuclei (superheat 0 °C).

References
5 Microstructure and properties of cold rolled strip

5.1 Introduction

DC cast AlMnMg alloy is commonly used for the fabrication of beverage cans due to its combination of good forming properties and good strength [1]. DC cast ingots are homogenised, hot rolled to a thickness of approximately ~ 2.5 mm [2], recrystallised during coiling and afterwards cold rolled to a thickness of ~ 0.3 mm. Strip casting is a competitive production process because less production steps are required to obtain the same final thickness. Strip cast material is cold rolled and finally recrystallised to obtain the desired formability properties [3, 4]. It is known that strip casting results in higher cooling rates [5] which gives smaller initial grain sizes and intermetallic phases and a higher solid solution level than DC cast material. These aspects influence the final properties [6].

Cold rolling of material results, for DC material as well as for strip cast material, in elongated pan cake grains. During rolling texture develops which is described in Chapter 6. The microstructural effects of rolling are that large intermetallic particles break up and particles become aligned with the rolling direction. It is found that strip cast material precipitates during cold rolling [7]. Recrystallisation of the material depends on many aspects. It is found that the subgrain size and grain size after rolling influences the recrystallisation behaviour in conventionally produced aluminium alloys [8, 9]. A correlation exists between the development of texture and subgrain size and depends strongly on the processing route. Solid solution, intermetallic phases, and distribution of phases influence recrystallisation [8-11]. In general, a high solid solution level increases the recrystallisation temperature and time [11], which is promoted by strip casting [5, 12]. Alloying elements such as Fe and Si have different effects on the recrystallisation behaviour of conventional cast material [2, 3, 13] but it is found that Mg in Al promotes recrystallisation [6]. Particles larger than 1 μm favour particle stimulated nucleation of new grains which is known as the PSN mechanism [14]. For DC cast material it is found that PSN occurs only at low temperatures. Inter-anneals for DC-cast material reduce the tendency of recrystallisation [15].

A general requirement for a formable alloy is a small grain size in the order of 10-30 μm [16]. It is known that large grain sizes cause surface damage during the forming operation which exhibits as orange peel. The formability of an alloy
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improves when a second phase dispersion exists of particles with a diameter of 0.5-1 μm and with an interparticle spacing of about 10 μm [16].

The goal of this study is to determine the effect of rolling on a strip cast and the effect of annealing on the rolled structure. The tensile properties of strip cast material are also determined. The development of grain size, particle size, and particle distribution after different treatments is discussed.

5.2 Experimental

AlMnMg alloys are strip cast on a copper wheel as described in Chapter 2 to obtain strip with a thickness of approximately 1 mm (Chapter 4). The melt is poured with a superheat of 40°C. The wheel temperature is 20°C. The alloys have different compositions based on the AA3004 alloy (Table 5.1).

Figure 5.1 shows the experimental procedure after casting. The strips are cold rolled to obtain a final thickness of approximately 170 μm which corresponds to a reduction of ~80%. The reduction is carried out in steps of 5%. There are two rolling conditions, that is rolling without inter-anneals and rolling with inter-anneals. The inter-anneals took place each time after an overall reduction of 20%. Annealing temperature and time are 345°C and 10 minutes which is a common treatment for AA3004 to recrystallise the material [17]. After rolling and rolling with inter-anneals, the strips are annealed at a temperature of 345°C for 10 minutes.

Table 5.1 Composition of the AlMnMg alloys; Al: bal

<table>
<thead>
<tr>
<th>Composition (wt, %)</th>
<th>Mn</th>
<th>Mg</th>
<th>Si</th>
<th>Fe</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>CP Al</td>
<td>0</td>
<td>~0.01</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>0.6</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>1.1</td>
<td>4.4</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>1.1</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>1.1</td>
<td>0.8</td>
<td>0.3</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>1.2</td>
<td>0.9</td>
<td>0.1</td>
<td>0.7</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>1.2</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>0.07</td>
</tr>
</tbody>
</table>
Microstructure and properties of cold rolled strip

![Diagram of the process](image)

Figure 5.1 Experimental procedure.

Both optical microscopy and tensile testing are applied to study the material obtained. From the rolled strip samples are taken from the plane parallel to the rolling direction and they are investigated on grain size and second phase characteristics.

Tensile tests are performed in accordance with the ASTM designation for very thin specimens [18]. From each strip at least three specimens were taken with the tensile direction in the rolling direction or in the transverse direction, that is 90° to the rolling direction. The dimensions of the used specimens are shown in Figure 5.2. The specimens were tested in a tensile machine with a constant displacement velocity of the crosshead of 1 mm/min. The maximum applied force was 1000 N and the load was applied until fracture of the specimen. From the tensile tests a stress-strain curve is obtained and the yield stress, tensile stress and strain to fracture are measured.

![Dimensions of tensile specimen](image)

Figure 5.2. Dimensions of tensile specimen.
5.3 Results

5.3.1 Grain size
Table 5.2 gives an overview of the grain size in the as-cast condition (Chapter 4) and in the rolled condition. Strongly deformed grains are difficult to etch and therefore the grains of some of the alloys could not be made visible which is shown by the empty cells in the table. The grains become elongated after rolling (Figure 5.3) and the thickness of the grains becomes in most cases smaller compared to the diameter of the grains in the as-cast condition.

After rolling and a final anneal the commercially pure (CP) Al shows fine recrystallised grains (Figure 5.3b) satisfying the requirements for a formable alloy [16]. The AlMn0.6Mg shows a smaller thickness and length of grains, than in the cold rolled condition, indicating that the grains are partly recrystallised. The grains of the AlMnMg+Si become coarser.

After rolling with inter-anneals the CP Al shows both spherical grains and elongated grains. The spherical grains have approximately the same diameter as the thickness of the elongated grains. The elongated grains are larger than after rolling only. Apparently, with inter-anneals grains become coarser and only some grains break up. Rolling with inter-anneals of AlMn0.6Mg does not influence the grain morphology and size when compared to rolling only. The recrystallisation effect observed after rolling and final annealing is not found here. Apparently rolling with inter-anneals leads to stress relief that suppresses recrystallisation, which is confirmed by [15].

The results indicate that inter-anneals have the same effect as a final anneal in case of only grain coarsening. For CP Al and AlMn0.6Mg, inter-anneals only coarsen grains and final annealing leads to recrystallisation. For the chosen annealing conditions DC-cast material is reported to recrystallise [12]. In our study only the CP Al recrystallises. The other alloys partly or do not recrystallise which means that recrystallisation is suppressed by strip casting. These findings are confirmed by [5]. The higher solid solution level in strip cast material compared to DC-cast material raises the recrystallisation temperature [12].
Table 5.2. Grain size development of the as-cast grain after different treatments. Dimensions of elongated (E) and spherical (S) grains. E: length x thickness.

<table>
<thead>
<tr>
<th></th>
<th>As-cast diameter (µm)</th>
<th>Rolled* (µm)</th>
<th>Rolled and final anneal (µm)</th>
<th>Rolled and inter-anneals** (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CP Al</td>
<td>S: 190</td>
<td>E: 682x120</td>
<td>S: 13</td>
<td>E: 1700x230, S: 171</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>S: 182</td>
<td>E: 1400x220</td>
<td>E: 860x125</td>
<td>E: 1400x220</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>S: 108</td>
<td>E: 500x44</td>
<td></td>
<td>E: 450x60</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>S: 170</td>
<td></td>
<td>E: 630x200</td>
<td></td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>S: 94</td>
<td>E: 190x54</td>
<td>E: 420x110</td>
<td>E: 420x110</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>S: 188</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>S: 125</td>
<td></td>
<td></td>
<td>E: 810x140</td>
</tr>
</tbody>
</table>

* 80% rolling reduction in steps of 5% reduction.
**inter-anneals after every overall reduction of 20%.

![Image](image-url)

Figure 5.3. The grain structure of the CP Al. a) 80% cold reduction. b) 80% cold reduction followed by a final anneal (recrystallised).

5.3.2 Intermetallic phases after casting and cold rolling

Table 5.3 shows morphology, size, and inter-particle spacing of the intermetallics after casting. The morphologies found for the as-cast condition are earlier presented in Chapter 4. The size of the intermetallics for the various morphologies is determined as follows: the length is measured in case of Chinese script, plates, and needles, and the diameter is measured in case of spheres. It should be noted that the Chinese script is difficult to measure due to the complexity of the shape. The average distance between similar particles is also given. Some Chinese script decorates clearly the boundary of dendrites.

It is seen from Table 5.3 that the as-cast condition gives small and large phases. The large phases correspond with a Chinese script morphology and the small
phases with a plate or needle morphology. The alloys can be divided in three categories based on their morphology. First, Chinese script is clearly seen in the CP Al. Second, Chinese script and other morphologies are seen in the AlMn0.6Mg, AlMnMg, AlMnMg+Si, and AlMnMg+Cr alloys. Third, only needles or plates are seen in the AlMnMg+Fe and AlMnMg4 alloys, respectively. The typical thickness of the Chinese script or needle is 0.5 μm. The smallest inter-particle spacing is found for the needle morphology and the inter-particle spacing of the plate morphology is almost similar for the different alloys and much larger than for the needle morphology. It seems that a refinement of phases appears simultaneously with the disappearance of the Chinese script.

Table 5.3. Characteristics of the intermetallics after casting. Symbols: ξ: Chinese script; □: plates; I: needles; О: spheres.

<table>
<thead>
<tr>
<th>morphology</th>
<th>size (μm)</th>
<th>inter-particle spacing (μm)</th>
<th>remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>ξ</td>
<td>20</td>
<td>33</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td></td>
<td>15</td>
<td>55</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>□</td>
<td>2</td>
<td>18</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>ξ</td>
<td>15</td>
<td>20</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td></td>
<td>13</td>
<td>44</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>I</td>
<td>3</td>
<td>7</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>ξ</td>
<td>10</td>
<td>22</td>
</tr>
</tbody>
</table>

Table 5.4 shows the morphology, the size, and the inter-particle spacing of the intermetallics after cold rolling. Some morphologies found for the rolled condition are given in Figure 5.4. It is found that almost all Chinese script is broken-up into small needles. The AlMn0.6Mg alloy shows a clustering of needles very similar to the Chinese script morphology. Also precipitation has occurred because three alloys show a smaller inter-particle spacing (Al, AlMnMg+Si, AlMnMg+Cr) and two alloys show larger phases (AlMnMg, AlMnMg4) due to the plate morphology. Rolling does not influence the AlMnMg+Fe alloy. Differences in morphology, size and inter-particle spacing between the alloys added with Si, Fe, and Cr disappear after rolling.
Table 5.4. Characteristics of the intermetallics after cold rolling. Symbols: ξ: Chinese script; □: plates ; I: needles; O: spheres.

<table>
<thead>
<tr>
<th></th>
<th>morphology</th>
<th>size (μm)</th>
<th>inter-particle spacing (μm)</th>
<th>remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>I</td>
<td>5</td>
<td>15</td>
<td>dispersed</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>I, O</td>
<td>7, 2</td>
<td>33, 18</td>
<td>clusters of I very similar to ξ which are aligned</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>□</td>
<td>4</td>
<td>33</td>
<td>aligned</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>□</td>
<td>5</td>
<td>33</td>
<td>clusters and aligned</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>I</td>
<td>2</td>
<td>7</td>
<td>dispersed</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>I</td>
<td>3</td>
<td>7</td>
<td>dispersed</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>I</td>
<td>2</td>
<td>7</td>
<td>dispersed</td>
</tr>
</tbody>
</table>

Figure 5.4. Intermetallic phases after cold rolling. a) Plates: AlMnMg4 alloy; b) Needles: AlMnMg+Si alloy; c) Clustered needles (Chinese script): AlMn0.6Mg alloy.
5.3.3 Intermetallic phases after final annealing

Table 5.5 shows the morphology, the size, and the inter-particle spacing of the intermetallics after final annealing. The alignment of intermetallics, which were aligned during rolling, does not change after the heat treatment, see Figure 5.5. This observation confirms that the material did not recrystallise although the size of the phases in most alloys after rolling is > 1 µm which could have lead to the PSN effect [14].

<table>
<thead>
<tr>
<th></th>
<th>morphology</th>
<th>size (µm)</th>
<th>inter-particle spacing (µm)</th>
<th>remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>☐</td>
<td>2</td>
<td>22</td>
<td>☐: clusters of I dispersed</td>
</tr>
<tr>
<td></td>
<td>☐</td>
<td>1</td>
<td>13</td>
<td></td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>☐</td>
<td>7</td>
<td>33</td>
<td>clusters of I and aligned</td>
</tr>
<tr>
<td></td>
<td>☐</td>
<td>2</td>
<td>18</td>
<td></td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>☐</td>
<td>5</td>
<td>11</td>
<td>clusters at grain boundaries;</td>
</tr>
<tr>
<td></td>
<td>☐</td>
<td></td>
<td></td>
<td>aligned</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>☐</td>
<td>5</td>
<td>11</td>
<td>slightly aligned</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>☐</td>
<td>2</td>
<td>7</td>
<td>dispersed</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>☐</td>
<td>2</td>
<td>7</td>
<td>dispersed</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>☐</td>
<td>3</td>
<td>7</td>
<td>dispersed</td>
</tr>
</tbody>
</table>

The morphology of only one alloy differs after a final anneal with respect to the cold rolling condition, namely the CP Al. The CP Al shows after final annealing Chinese script instead of needles. This Chinese script consists of clustered needles similar to the features seen in Figure 5.4c. Apparently the needles that were already present after cold rolling are clustered due to recrystallisation of the CP Al. These clusters of needles are positioned at the grain boundary because the interparticle spacing is similar to the recrystallised grain size.

The size and the inter-particle spacing of the intermetallics does not change in most alloys except in the AlMnMg and AlMnMg4 alloys. In the AlMnMg4 alloy the size becomes smaller but the inter-particle spacing does not change. Apparently particles partly dissolve during annealing. In case of the AlMnMg alloy the size does not change but inter-particle spacing becomes smaller and the particles are more dispersed.
Microstructure and properties of cold rolled strip

![Microstructure images](image)

Figure 5.5. Microstructure of AlMnMg4 alloy aligned (a) and AlMnMg+Si alloy dispersed (b) after rolling and final anneal.

### 5.3.4 Intermetallic phases after cold rolling with inter-anneals

Table 5.6 shows the morphology, the size, and the inter-particle spacing of the intermetallics after cold rolling with inter-anneals.

<table>
<thead>
<tr>
<th></th>
<th>morphology</th>
<th>size (μm)</th>
<th>inter-particle spacing (μm)</th>
<th>remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>I; O</td>
<td>2, 1</td>
<td>170, 13</td>
<td>I: on grain boundary; O: dispersed</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>I; O</td>
<td>7, 2</td>
<td>33, 18</td>
<td>clusters and aligned</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>□</td>
<td>2</td>
<td>13</td>
<td>clusters and aligned</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>□</td>
<td>2</td>
<td>11</td>
<td>aligned</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>□</td>
<td>3</td>
<td>14</td>
<td>dispersed</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>I</td>
<td>2</td>
<td>7</td>
<td>aligned</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>□</td>
<td>6</td>
<td>44</td>
<td>aligned</td>
</tr>
</tbody>
</table>

Comparing rolling and inter-anneals with the as-cast structure (Table 5.6 with Table 5.3) it is found that the Chinese script morphology does not appear. Apparently it has broken up due to rolling. Beside breaking up of phases, growth of phases occurs in the AlMnMg+Si and AlMnMg+Cr alloys. Smaller phases are found in the AlMnMg4 alloy, which indicate that partial dissolution of phases takes place which is also found after rolling and final annealing. The inter-anneals do not affect the AlMn0.6Mg and AlMnMg+Fe alloys since it shows the same microstructure as is seen after rolling only.

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The size and inter-particle spacing of the intermetallics after inter-anneals differ from the rolled condition for most alloys. Larger phases are seen in the Al, the AlMnMg+Si and the AlMnMg+Cr alloys as seen in Figure 5.6. These phases are also larger than after rolling and final annealing. In contrast, smaller phases are seen in the AlMnMg4 and the AlMnMg alloys, compare Figure 5.7 with Figure 5.4a. These phases are also smaller than after rolling and final annealing.

![Image](image1.png)

**Figure 5.6.** Intermetallic phases of AlMnMg+Cr after rolling (a) and rolling combined with inter-anneals (b).

![Image](image2.png)

**Figure 5.7.** Intermetallic phases of AlMnMg4 after rolling with inter-anneals.

5.3.5 **Tensile properties**

In Figure 5.8 the ultimate tensile stress (UTS) of the strip cast material is plotted against the strain to fracture for most alloys in either rolled condition only or rolled with inter-anneals. It is found that stress decreases and strain increases with applying the inter-anneals. It is seen that the strength properties can be divided into 3 categories, namely high strength which applies to the AlMnMg4 alloy, low strength which applies to the CP Al, and intermediate strength which applies to the remaining
Microstructure and properties of cold rolled strip

alloys. According to literature [11] the effect of the two main alloying elements of aluminium, Mn and Mg, is to increase the strength by solid solution hardening. Due to the high cooling rates it is expected that all Mg is in solution which gives AlMnMg4 the largest strength. CP Al has the smallest strength. The remaining alloys have an intermediate composition and are therefore expected to have an intermediate strength. Concerning the strain to fracture it is found that for the CP Al premature failure occurred. Due to coarsening, grain size becomes comparable with specimen thickness.

Figure 5.8. Ultimate tensile stress versus strain to fracture for strip cast material in the rolled condition (■: AlMnMg+Si, AlMnMg+Fe, AlMnMg+Cr, AlMn0.6Mg; ●: AlMnMg4) and in the rolled condition with inter-anneals (◇: Al; □: AlMnMg+Si, AlMnMg+Fe, AlMnMg+Cr, AlMn0.6Mg; ○: AlMnMg4).

Figure 5.9. Ultimate tensile stress versus strain to fracture for the rolled condition with inter-anneals. The rolling direction (◇: Al; □: AlMnMg+Fe; △: AlMnMg+Cr) and transverse direction (◆: Al; ■: AlMnMg+Fe; ▲: AlMnMg+Cr).

Figure 5.9 shows the tensile behaviour for some alloys in the rolling direction and in the transverse direction. It is seen that with the transverse direction
the strain to fracture decreases while the strength remains constant. This effect is similar to conventional material.

5.4 Discussion

5.4.1 Effects of rolling, annealing and inter-anneals on intermetallics.

It is found that large intermetallic phases ($\geq 13 \text{ m}$) resembling a Chinese script morphology in as-cast strip have broken up after rolling. The phases that broke up become aligned after rolling. During solidification a large amount of Mn, Fe and Si remain in solid solution due to the high cooling rate. During rolling, in some alloys this supersaturation is partly taken away by precipitation. It is known that solubility of Fe and Si in aluminium at room temperature is low and therefore their presence promotes precipitation of $\text{Al}_{12}(\text{MnFe})_3\text{Si}_2$ phases [11]. On one hand, precipitation during rolling leads to new precipitates ($\text{AlMnMg+Si, AlMnMg+Cr}$). This is also observed in the CP Al which is attributed to impurities of Fe and Si. On the other hand, precipitation leads to growth of existing phases ($\text{AlMnMg, AlMnMg4}$). Apparently additions of Si, Fe, or Cr to the AlMnMg alloy hinder growth of existing phases and therefore promote precipitation of new phases in case of Si or Cr additions. An addition of Fe does not induce precipitation.

Final annealing does not have much effect on the intermetallics obtained after rolling, except for the Al and AlMnMg4 alloy. In case of the CP Al recrystallisation takes place, leading to clustering of needle-shaped phases. In the AlMnMg4 alloy, dissolution of precipitates is observed which is according to what is expected from the phase diagram.

Rolling with inter-anneals also results in breaking up of large phases. The phases that broke up become also aligned. Rolling with inter-anneals suppresses the precipitation, which was found after rolling only. New precipitates are not found and also growth of existing phases in AlMnMg and AlMnMg4 is not observed. Therefore, it is concluded that precipitation during rolling only occurs if the accumulated strain builds up a sufficiently high driving force to induce precipitation from the saturated solid solution. Inter-anneals interrupt and reduce the accumulated strain which lead to an insufficient driving force. It is observed in case of AlMnMg+Si and AlMnMg+Cr alloys that growth of phases occurs in combination with dissolution of smaller phases. Apparently, in these alloys Ostwald ripening take place [11].
5.4.2 Strength properties

Figure 5.10 shows the tensile properties of the strip cast material compared to those of a conventional produced AA3004 alloy which has been cold rolled to a thickness of 0.17 mm. A trend is seen of decreasing strength with increasing strain to fracture for the AlMnMg+Si, AlMnMg+Fe, AlMnMg+Cr, and AlMn0.6Mg alloys. Since the same trend line can be imposed on the data points of AlMnMg4, it is assumed that the same trend line can be applied to CP Al.

![Graph showing ultimate tensile stress versus strain to fracture for strip cast material after rolling and after inter-anneals.](image)

**Figure 5.10.** Ultimate tensile stress versus strain to fracture for strip cast material after rolling (■: AlMnMg+Si, AlMnMg+Fe, AlMnMg+Cr, AlMn0.6Mg; ○: AlMnMg4) and after rolling with inter-anneals (◇: Al; □: AlMnMg+Si, AlMnMg+Fe, AlMnMg+Cr, AlMn0.6Mg; ○: AlMnMg4).

![Graph showing mechanical properties of strip cast material after rolling of AlMn0.6Mg, AlMnMg+Si, AlMnMg+Fe, or AlMnMg+Cr compared to typical tensile properties of AA3004.](image)

**Figure 5.11.** Mechanical properties of strip cast material after rolling of AlMn0.6Mg, AlMnMg+Si, AlMnMg+Fe, or AlMnMg+Cr (□) compared to typical tensile properties of AA3004 (----- [19]) and DC cast material of 0.17 mm thickness (◇).
Figure 5.11 shows the tensile properties of the strip cast material (AlMn0.6Mg, AlMnMg+Si, AlMnMg+Fe, AlMnMg+Cr) after rolling compared to a conventionally produced AA3004 alloy, which has a comparable thickness (0.17 mm) and compared to the ASTM standard for AA3004. It is found that the strip cast material after rolling satisfies the property limits of the ASTM standard which is in agreement with [5]. The strip cast material shows lower values of mechanical properties than the DC cast material with a similar thickness. There are two possible explanations for the lower properties of the strip cast material. One explanation is that the as-cast strip has fewer grains in the thickness of the strip than DC cast material since this latter material is hot rolled prior to cold rolling. Since the strip shows after rolling elongated grains (width of ~155 um) and the thickness of the grain correspond to the thickness of the strip (~170 um), the grain boundary dominates the strain to fracture. Smaller and more grains obtained by a heat treatment leading to recrystallisation, should give a larger strain to fracture. Second explanation is that the strip cast material was not of optimal quality because of technical limitations in the laboratory equipment. Optimalised strip could therefore provide mechanical properties which exceeds the ASTM standard.

5.5 Conclusions

- Rolling of strip cast material:
  - results in elongated grains and aligned intermetallics similar to that of conventional cast material,
  - breaking up of the Chinese script morphology,
  - gives intermetallics with a size of 2-7 μm,
  - an inter-particle spacing and morphology that depends largely on alloy composition,
  - with minor additions of Si, Fe, or Cr to the AlMnMg alloy hinders growth of intermetallics during rolling,
  - gives mechanical properties within the ASTM standard (UTS 300-450 MPa, strain to fracture of 2%).

- Final annealing of rolled strip cast material:
  - results in recrystallised grains for CP Al,
  - does not influence the morphology, size and inter-particle spacing of rolled strip cast material.
Microstructure and properties of cold rolled strip

- Recrystallisation is hindered in rolled strip cast material due to a higher amount of alloying elements in solid solution.
- Inter-annealing of rolled strip cast material:
  - does not influence the grain size of rolled strip cast material,
  - affects the size and inter-particle spacing of most studied alloys,
  - suppresses precipitation during rolling.

References
18. ASTM designation E8-69.
6 Texture in strip cast material

6.1 Introduction

The process route from DC cast slab to rolled sheet with a thickness <0.3 mm for the deep drawing process for beverage cans is given in Figure 6.1.

![Diagram showing process route from DC cast slab to rolled sheet](image)

**Figure 6.1.** The process route from DC cast slab to rolled sheet for deep drawing. After each step the corresponding texture is given. $T_{exit}$: exit temperature of the hot rolled sheet. $RX$: volume percentage recrystallisation texture.
Strip Casting of Aluminium Alloys

In the deep drawing process of beverage cans from circular blanks, the wall of the final cup can vary in height. This general phenomenon is called earing [1-6] and it is caused by anisotropic material. Since the material in the flange is stretched in one direction (radially) and compressed in the perpendicular direction (circumferentially) drawing requires a material that is able to resist thinning. This feature is expressed by the plastic strain ratio, the $r$ value, and it is defined as the ratio of the true width strain to the true thickness strain in the uniform elongation region of a tensile test [1]. High $r$ values mean good resistance against thinning of the material. For anisotropic material the $r$ value changes with the tensile direction in the sheet and therefore usually the average of three directions is taken which is expressed by $r_m$:

$$r_m = \frac{r_0 + r_{45} + r_{90}}{4} \quad (6.1)$$

where the subscript refers to the angle with respect to the rolling direction. For Al alloys, $r_m$ values range from 0.6 to 0.8 and for high formable steels values of 2.8 are found [1]. The planar anisotropy, $\Delta r$, is used to determine the difference between the $r$ values:

$$\Delta r = \frac{r_0 - 2r_{45} + r_{90}}{2} \quad (6.2)$$

No earing is found when this value is zero. Optimum drawability is obtained when the combination of $r_m$ and $\Delta r$ values are respectively high and low.

Isotropic material will not give earing. In this situation, the material has a random texture which means that no particular crystal orientation is present. In contrast, in anisotropic material two types of texture can be distinguished beside the random texture. Each of them leads to a certain direction in which earing occurs. A deformation texture is formed after cold rolling. The deformation texture consists of Copper, S and Brass orientations and those orientations are also called the $\beta$-fibre. These orientations are expressed in a 2D pole figure such as shown schematically in Figure 6.2a. Deformation texture components lead to 45° earing with respect to the rolling direction. A recrystallisation texture is formed after recrystallisation, and include Cube, Goss, R, P, and Q orientations. The R orientation is similar to the S orientation and refers to a retained S orientation after a recrystallisation heat treatment. Figure 6.2 shows schematically the positions of the Cube and Goss orientations in a pole figure. Recrystallisation texture components lead to 0°/90° earing with respect to the rolling direction. Both types result in four ears. A
combination of the deformation texture components and recrystallisation texture components lead to six (0°/180°/45°) or eight (0°/90°/45°) ears. Whether six or eight ears appear, depends on the strength of the Cube component [5]. The influence of P and Q orientations on earing is not mentioned in literature.

Methods to achieve a random texture such as imposing small rolling reductions with frequent intermediate annealing, or multidirectional rolling are expensive [7]. Therefore usually a balanced texture [8] between the recrystallisation components and the deformation components is considered as optimal. In conventional material the deformation components are compensated with recrystallisation components by starting with Cube and Goss texture components prior to cold rolling [7, 9, 10]. During cold rolling part of the Cube and Goss orientations are rotating towards deformation orientations. The Cube component rotates slower than other components. To obtain minimal earing in 80-90% cold rolled conventional material, a balanced texture is in general achieved with 20-45% recrystallisation texture components prior to cold rolling [5, 7, 9, 10]. A final heat treatment before deep drawing gives the further necessary compensating recrystallisation texture components.

In strip cast material the as-cast texture is less strong than in conventionally cast material, such as Direct Chill cast material [11]. This is explained by the high cooling rates in combination with the thin strip that results into a larger random component. This means that also the Cube orientations in as-cast strip are weaker prior to cold rolling. This feature makes it complicated to compensate the deformation texture during cold rolling. A final heat treatment that results in recrystallisation of the strip can only compensate the deformation texture if PSN appears that favours the formation of a random component instead of recrystallisation components. The recrystallisation mechanism that takes place is influenced by solid solution level, second phase dimensions, and composition [11].

The aim of this study is to research the effect of heat treatments and alloying elements on the texture in aluminium strip cast material. The results are compared to conventionally produced material.

6.2 Pole figures

In Figure 6.2 the positions of the deformation and recrystallisation texture components are schematically represented in a {111} pole figure. The vertical axis
of the pole figure is the rolling direction (RD) and the horizontal axis is the transverse direction (TD), which is the direction perpendicular to the rolling direction.

![Diagram of pole figures](image)

**Figure 6.2. Schematical representation of the positions of main texture components in a {111} pole figure. a) Deformation texture components: Brass (B), S, and Copper (C). b) Recrystallisation texture components: Cube and Goss (G). RD = rolling direction; TD = transverse direction.**

### 6.3 Experimental

Texture measurements are performed on strip cast and Direct Chill (DC) cast AlMnMg based alloys. Strip cast material is produced on a single-roll strip caster as described in Chapter 2. The contents of alloying elements of the used strip cast alloys are given in Table 6.1. The DC cast material, which is a commercial AlMnMg alloy (Table 6.2), was cast at Corus Aluminium Duffel (Belgium) where it was hot rolled to 4 mm thickness. This is the delivered condition which is used as reference material.
Table 6.1 Composition of the strip cast AlMnMg alloys; Al: bal.

<table>
<thead>
<tr>
<th>Composition (wt. %)</th>
<th>Mn</th>
<th>Mg</th>
<th>Si</th>
<th>Fe</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>CP Al</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMn0.6Mg</td>
<td>0.6</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg4</td>
<td>1.1</td>
<td>4.4</td>
<td>0.1</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg</td>
<td>1.0</td>
<td>1.0</td>
<td>0.3</td>
<td>0.4</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Si</td>
<td>1.1</td>
<td>0.8</td>
<td>0.3</td>
<td>0.2</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Fe</td>
<td>1.2</td>
<td>0.9</td>
<td>0.1</td>
<td>0.7</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>AlMnMg+Cr</td>
<td>1.2</td>
<td>1.1</td>
<td>0.1</td>
<td>0.2</td>
<td>0.07</td>
</tr>
</tbody>
</table>

Table 6.2. Composition of the DC cast AlMnMg alloy; Al: bal.

<table>
<thead>
<tr>
<th></th>
<th>Mn</th>
<th>Mg</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>AlMnMg</td>
<td>1.0</td>
<td>1.0</td>
<td>0.2</td>
<td>0.4</td>
<td>0.2</td>
</tr>
</tbody>
</table>

The process routes for both strip cast material and the as-received DC cast material are given in Figure 6.3. Material is cold rolled to a reduction of approximately 80% in steps of 5% reduction. The inter-anneals are performed after every overall reduction of 20% and are intended to retard the development of deformation orientations. After the final heat treatment the samples are air-cooled to room temperature. The final heat treatment is intended to recrystallise the material and to obtain recrystallisation orientations or random orientations.

Texture measurements are carried out by Corus (The Netherlands), and are indicated with a ‘T’ in Figure 6.3. The texture is measured on an area of 50 mm x 50 mm in the centre of the strip thickness. In addition, for two cases the texture close to the wheel surface is measured, that is in the as-cast condition and after 80% reduction, to study texture variation through the thickness of the strip cast material. The texture is measured by a standard X-ray goniometer. Four incomplete pole figures {111}, {220}, {200}, and {311} are measured. From the pole-figure data, Orientation Distribution Functions (ODFs) are determined using the software package MTM-FHM developed by Van Houtte [12], University of Leuven, Belgium. The software uses the standard series expansion method of Bunge [13]. The three-dimensional ODFs are as usual graphically represented by sections through the $\varphi_2$ direction in steps of 5 degrees whereas in this study only $\varphi_2=0^\circ$ sections will be given. Volume percentages of texture components are derived from the ODF data with a spread of 11° around specific texture orientations. The percentage deformation texture components is taken as the sum of the Brass,
Copper, and S texture components, whereas the volume percentage recrystallisation texture components is taken as the sum of Goss, and all Cube components. The P and Q orientations are presented separately. The complete pole figures are obtained from the ODF calculations. In this paper only the \{111\} pole figures will be given since these pole figures show the necessary information about the orientations.

The \( r \) values of the sheet are calculated from the ODF data using the same software by Van Houtte and assuming slip in the \{111\} direction [12]. From these \( r \) values the \( r_m \) and \( \Delta r \) are calculated.

*Figure 6.3. Process routes for strip cast material and as-received DC cast material. Texture measurements are indicated with T.*
6.4 Results

6.4.1 Texture in strip cast AlMnMg alloy

Figure 6.4 shows the \{111\}-pole figures of strip cast material after casting and after cold rolling according to Figure 6.3.

![Texture figures]

Figure 6.4. \{111\}-pole figures of strip cast material. a) as-cast; b) as-cast and homogenised; c) as-cast and 80% cold rolled; d) as-cast, homogenised and 80% cold rolled. Minimum and maximum levels are indicated with a number. Rolling direction is vertical.
It is seen that after casting and homogenisation a random texture is found. The maximum intensity is 1.3 times random. After 80% cold reduction, a typical deformation texture with β-fibre components is seen for both conditions. The maximum intensity found here is 5.0 times random which means that the texture becomes stronger due to rolling. The volume percentage deformation texture components and recrystallisation texture components are given in Figure 6.5. It is seen that the random texture found in Figure 6.4 for as-cast strip consist of approximately 10% volume percentage deformation texture components and 10% recrystallisation texture components present in as-cast strip. This percentage does not show to affect the random texture. It is seen that the volume percentages are not influenced by the homogenisation treatment. Further, the texture in the centre of the as-cast strip does not differ from the texture found at the surface of the as-cast strip.

![Figure 6.5. Volume percentages texture components near the surface (0.25 t) and in the centre (0.5t) of strip cast and rolled AlMnMg. H=homogenised strip. □: deformation texture components; ■: recrystallisation texture components; ▲: random.](image)

![Figure 6.6. Volume percentage deformation texture components (Copper, Brass, R/S) near the surface (□) and in the centre of rolled strip (■) (AlMnMg).](image)
Figure 6.6 shows the volume percentage Copper, Brass, and R/S components near the surface and in the centre of rolled strip. It is seen that the volume percentage R/S remains constant. The volume percentage Brass decreases from the surface towards the centre whereas the volume percentage Copper increases from the surface towards the centre of the strip.

6.4.2 Effect of final heat treatment
Figure 6.7a shows the {111}-pole figure of strip cast material after cold rolling and a final heat treatment of 480 °C. The strong deformation texture as was seen after cold rolling is strongly reduced by this treatment. Figure 6.7a shows a balanced texture of deformation texture components and recrystallisation texture components. The maximum intensity is 1.6 times random. In addition, for the homogenised condition, which was rolled and heat treated at 480 °C (Figure 6.7b), a clear but weak recrystallisation texture is found with a maximum intensity of 1.3 times random. The microstructures of both treatments show fully recrystallised grains.

Figures Figure 6.7b-d show the effect of recrystallisation temperature on the texture in homogenised strip cast material. The maximum intensity for all temperatures in the homogenised condition is 1.3 times random. After a final anneal of 345 °C mixed texture components are seen that consist of recrystallisation and deformation texture components. The recrystallisation texture becomes more clear at a temperature of 415 °C. The texture remains more or less constant when the temperature is increased to 480 °C. The microstructures of these materials after the final heat treatments show fully recrystallised grains.
Figure 6.7. [111]-pole figures of strip cast material after cold rolling and a final heat treatment. a) strip cast material, cold rolled and heat treated at 480 °C; homogenised strip cast material, cold rolled and heat treated at different temperatures: b) 480 °C; c) 415 °C; d) 345 °C. Minimum and maximum levels are indicated with a number. Rolling direction is vertical.

Figure 6.8 shows the volume percentages of the deformation texture components and the recrystallisation texture components in strip cast material after a final heat treatment. Although the pole figures differ from each other, it is found that the volume percentages do not differ much. The volume percentages deformation texture and recrystallisation texture are similar. This corresponds with the weak recrystallisation texture or balanced texture found in the pole figures.
Figure 6.8. Volume percentage texture components in strip cast AlMnMg after rolling and a final heat treatment (HT) and homogenised (H) strip cast AlMnMg after rolling and final heat treatments. □: deformation texture components; ■: recrystallisation texture components; ★: random.

Figure 6.9 shows a detailed view of the volume percentages texture components of strip cast AlMnMg and strip cast AlMnMg that is rolled and heat treated. It is found that these volume percentages are nearly the same. Further it is found that the volume percentage P and Q components are not influenced by homogenising, rolling, and heat treatments and are 5.7±0.2% and 2.2±0.9% respectively.

Figure 6.9. Volume percentages of texture components of strip cast AlMnMg after strip casting (□) and after a final heat treatment at a temperature of 480 °C (■). ND: normal direction; RD: rolling direction; R: rotated.

From the former sections the microstructure and texture of the AlMnMg alloy for the different treatments can be summarised and are presented in Table 6.3.
### Table 6.3. Texture and microstructure in strip cast AlMnMg.

<table>
<thead>
<tr>
<th></th>
<th>texture</th>
<th>microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-cast</td>
<td>random</td>
<td>equiaxed grains</td>
</tr>
<tr>
<td>As-cast + homogenised</td>
<td>random</td>
<td>equiaxed grains</td>
</tr>
<tr>
<td>Rolled</td>
<td>deformation</td>
<td>elongated grains</td>
</tr>
<tr>
<td>Final treatment</td>
<td>balanced</td>
<td>recrystallised grains</td>
</tr>
</tbody>
</table>

#### 6.4.3 Effect of alloying elements on texture in strip cast material

The textures after rolling of the different aluminium based alloys (Table 6.1) are all showing deformation textures with maximum intensities of approximately 4.0 times random. Figure 6.10 shows the volume percentages of deformation texture components and the recrystallisation texture components in the rolled strip. It is found that the AlMnMg+Fe alloy and the AlMnMg4 have the lowest and highest volume percentage deformation texture components respectively. The volume percentages of the recrystallisation texture components show little variation between the alloys.

![Figure 6.10. Volume percentages of texture components after cold rolling for AlMnMg based alloys.](image)

Figure 6.10. Volume percentages of texture components after cold rolling for AlMnMg based alloys. ☐: deformation texture components; ■: recrystallisation texture components; ☐: random.

Figure 6.11 shows the volume percentages of the Copper, Brass, and R/S components of the rolled strip. It is seen that these components differ for each alloy and in particular for the Al alloy and the AlMnMg4 alloy. It is seen that only in the Al and AlMnMg+Fe alloys the Copper component is larger than the Brass component. Therefore, the observation from Figure 6.6 that the Copper component is larger than the Brass, strongly depends on the alloy composition.
Texture in strip cast material

Figure 6.11. Volume percentage deformation texture components of rolled AlMnMg based alloys. □: Al; ■: AlMnMg+Si; ★: AlMnMg+Fe; ●: AlMnMg+Cr; ◊: AlMnMg4.

The textures after rolling and a final heat treatment for the different alloys are all typical deformation textures. Figure 6.12 shows the volume percentages of deformation texture components and recrystallisation texture components for the different alloys after cold rolling and a final heat treatment.

Figure 6.12. Volume percentages texture components after cold rolling and a final heat treatment. □: deformation texture components; ■: recrystallisation texture components; ★: random.

Comparing Figure 6.12 with Figure 6.10, it is found that the volume percentage deformation texture after the final heat treatment changes only in the AlMnMg4 alloy where the volume percentage deformation texture decreases from 44% to 33% and the volume percentage recrystallisation texture increases from 5% to 9%.

6.4.4 Effect of inter-anneals on the texture in strip cast material
The CP Al shows after rolling with inter-anneals a recrystallisation texture similar to that seen in Figure 6.7b but with a somewhat larger maximum intensity of 1.6 times random. The remaining alloys show typical deformation textures with maximum
intensities of 5 times random. Figure 6.13 shows the volume percentages of deformation texture components and recrystallisation texture. The volume percentages are similar to those after cold rolling (apart from CP Al), see Figure 6.10. For the CP Al it is seen that, with inter-anneals, the volume percentage deformation texture decreases and the recrystallisation texture increases, in line with the difference in the pole figures. Since in this alloy the random component remains constant compared to rolling only, the decrease of deformation texture components equals the increase of recrystallisation texture components. It appears that the recrystallisation mechanism in CP Al is totally in favour of the development of recrystallisation texture components. Further it is seen in the CP Al that the volume percentages deformation texture and recrystallisation texture are similar, which corresponds to the mixed texture components found in the pole figure.

![Bar chart showing volume percentages of texture components](image)

Figure 6.13. Volume percentages texture components after cold rolling with inter-anneals. □: deformation texture components; ■: recrystallisation texture components; ◯: random.

The texture and microstructure (Chapter 5) found for the AlMnMg based alloys after the different treatments are summarised in Table 6.4.

<table>
<thead>
<tr>
<th>Treatment</th>
<th>Texture</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>As cast</td>
<td>random</td>
<td>equiaxed grains</td>
</tr>
<tr>
<td>Rolled</td>
<td>deformation</td>
<td>elongated grains</td>
</tr>
<tr>
<td>Rolled with inter-anneals</td>
<td></td>
<td></td>
</tr>
<tr>
<td>CP Al</td>
<td>balanced</td>
<td>partly recrystallised grains</td>
</tr>
<tr>
<td>AlMnMg based alloys</td>
<td>deformation</td>
<td>elongated grains</td>
</tr>
<tr>
<td>Final heat treatment</td>
<td></td>
<td></td>
</tr>
<tr>
<td>CP Al</td>
<td>deformation</td>
<td>recrystallised grains</td>
</tr>
<tr>
<td>AlMnMg based alloys</td>
<td>deformation</td>
<td>elongated grains</td>
</tr>
</tbody>
</table>

100
6.4.5 *Anisotropy in AlMnMg strip cast material*

Figure 6.14 shows the $r$ values in AlMnMg strip cast material. It is seen that after casting there is some anisotropy in the strip which will exhibit as trogs at 0°, 45°, and 90°. Rolling results in an increase of the $r$ value at 45° due to the formation of Brass, S, and Copper components and a decrease of the $r$ value at 90° due to the decrease of Cube components. Therefore the rolled strip will give 45° earing which means four ears and since the $r$-value at 90° is somewhat lower than at 0°, it will give trogs at 90°. Final annealing diminish largely the 45° earing but it is still present. In contrast to homogenised strip cast material that shows less anisotropy in the 45° direction after final annealing.

![Graph](image-url)

*Figure 6.14. The r values in AlMnMg strip cast material. As-cast: ---; rolled: --- ---; final annealed (480 °C): ---; homogenised, rolled and final annealed (480 °C): ---.*

<table>
<thead>
<tr>
<th></th>
<th>$r_m$</th>
<th>$\Delta r$</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-cast</td>
<td>0.9</td>
<td>0.01</td>
</tr>
<tr>
<td>Rolled</td>
<td>1.5</td>
<td>-1.7</td>
</tr>
<tr>
<td>Final annealed (480 °C)</td>
<td>0.9</td>
<td>-0.4</td>
</tr>
<tr>
<td>Homogenised, rolled, final annealed (480 °C)</td>
<td>0.8</td>
<td>-0.1</td>
</tr>
</tbody>
</table>

Table 6.5. The average $r$ value, $r_m$, and the planar anisotropy, $\Delta r$, of strip cast AlMnMg material.

Table 6.5 shows the average $r$ value, $r_m$, and the planar anisotropy, $\Delta r$, of strip cast AlMnMg material. As was expected the as-cast strip gives the best combination between $r_m$ and $\Delta r$ value, with respect to the other conditions. The final annealed material shows $r_m$ values which are just above the standard values for aluminium.
alloys whereas the $\Delta r$ of the homogenized, rolled and final annealed strip will give a lower earing percentage than final annealing only.

6.4.6 Texture in DC cast material

Figure 6.15 shows the $\{111\}$ pole figures of as-received AlMnMg material which is 80 % cold rolled, as well as as-received AlMnMg material which is additionally annealed and 80 % cold rolled.

![Pole figures](image)

*Figure 6.15. $\{111\}$ pole figures of DC cast material. a) as-received; b) as-received and annealed, c) 80 % rolled; d) annealed and 80% rolled. Minimum and maximum levels are indicated with a number. Rolling direction is vertical.*
The as-received material was DC cast and hot rolled on industrial scale. The texture found in the as-received AlMnMg material is a typical deformation texture with a maximum intensity of 6.4 times random. This texture is stronger than the deformation texture found in strip cast material after lab scale rolling. The as-received and annealed AlMnMg material, Figure 6.15b, shows a strong recrystallisation texture with a maximum intensity of 4.0 times random. The microstructure shows recrystallised grains after the annealing treatment. Both materials show after cold rolling a deformation texture as can be seen in Figure 6.15c and d. The as-received material shows nearly the same deformation texture after rolling, although slightly more pronounced since the maximum intensity is now 8 times random. The pole figure found for the as-received material, that is annealed prior to rolling, shows a deformation texture which is less pronounced but that corresponds to rolled strip cast material, compare Figure 6.15d with Figure 6.4c.

Figure 6.16 shows the \{111\} pole figures of the rolled material after the final heat treatment. Final annealing gives for both conditions mixed texture components that consist of deformation and recrystallisation texture components. The microstructures show recrystallised grains. The as-received material that is cold rolled and final annealed, has a maximum intensity of 2.5 times random, whereas the material that is annealed prior to rolling shows after final annealing a somewhat lower maximum intensity.

Figure 6.16. \{111\} pole figures of AlMnMg DC cast material that is cold rolled and final annealed (a) or annealed prior to rolling and final annealed (b). Minimum and maximum levels are indicated with a number. Rolling direction is vertical.
Figure 6.17. Volume percentages texture components in AlMnMg material and in annealed AlMnMg material that is additionally 80% cold rolled and final annealed (FA) at 345 °C for 30 minutes. □: deformation texture components; ■: recrystallisation texture components; □□: random.

The volume percentages deformation texture components and recrystallisation texture components are given in Figure 6.17 for as-received material and annealed material that is additionally cold rolled and final annealed. The as-received material shows after rolling a high percentage deformation texture (50-60%). This is much stronger than the percentages deformation texture found in strip cast material (30-40%). Further, after final annealing, the deformation texture components strongly decrease and the recrystallisation texture components increase resulting in equal volume percentages. This corresponds with the mixed texture components found in the pole figure. Two differences are observed compared to strip cast material that is final annealed. First, the decrease of deformation texture components does not equal the increase of recrystallisation texture components as was seen in strip cast material. Second, the balanced texture is stronger than is seen in strip cast material, approximately 20% versus 10% for strip cast material.

Annealing the as-received material gives a strong decrease and increase of deformation texture and recrystallisation texture respectively. After rolling the deformation texture increases, as expected, and this increase roughly equals the decrease of the recrystallisation texture. These percentages texture components are comparable with the cold rolled strip cast material (see Figure 6.5). Final annealing shows a further decrease of the deformation texture components and the recrystallisation texture components remain similar. The deformation texture is somewhat stronger than the recrystallisation texture.

The textures and microstructures found for the DC cast AlMnMg material after the different treatments are summarised in Table 6.6.
Table 6.6. The texture and microstructure of DC cast AlMnMg material after the different treatments.

<table>
<thead>
<tr>
<th></th>
<th>Texture</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>As received</td>
<td>deformation</td>
<td>elongated grains</td>
</tr>
<tr>
<td>Annealed</td>
<td>recrystallisation</td>
<td>recrystallised grains</td>
</tr>
<tr>
<td>Rolled</td>
<td>deformation</td>
<td>elongated grains</td>
</tr>
<tr>
<td>Annealed, rolled, final annealed</td>
<td>balanced</td>
<td>recrystallised grains</td>
</tr>
<tr>
<td>Rolled, final annealed</td>
<td>balanced</td>
<td>recrystallised grains</td>
</tr>
</tbody>
</table>

Figure 6.18. The $r$ values in DC cast AlMnMg material after final annealing. Cold rolled and final annealed: —; annealed prior to cold rolling and then final annealed: ----.

Figure 6.18 shows the $r$ values of DC cast material after final annealing. For material without annealing prior to rolling the $r$ value at $0^\circ$ and $90^\circ$ is larger than at $45^\circ$. This material will give ears at $0^\circ$ and $90^\circ$. It is seen that material with previous annealing gives at $90^\circ$ a smaller $r$ value than at $0^\circ$ and $45^\circ$. This will result in a trog at $90^\circ$. The corresponding $r_m$ and $\Delta r$ values are given in Table 6.7.

Table 6.7. The average $r$ value, $r_m$, and the planar anisotropy, $\Delta r$, DC cast AlMnMg material after final annealing.

<table>
<thead>
<tr>
<th></th>
<th>$r_m$</th>
<th>$\Delta r$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cold rolled and final annealed</td>
<td>0.8</td>
<td>0.3</td>
</tr>
<tr>
<td>Annealed, rolled, and final annealed</td>
<td>0.8</td>
<td>-0.2</td>
</tr>
</tbody>
</table>
6.5 Discussion

6.5.1 Texture in as-cast strip
In this study, it is found that the as-cast texture is random in the centre of the strip. This is in agreement with the microstructure of the as-cast strip which shows equiaxed grains. The as-cast texture is also random near the wheel surface of the strip and the percentage texture components is similar to that in the centre of the strip. This is confirmed by the microstructure of the as-cast strip which shows at the wheel side also equiaxed grains although the dendrites inside the grains are directional. However, this does not lead to texture differences throughout the thickness.

In conventional DC cast ingots usually a random texture is present [15]. This is due to the added grain refiners in the melt, which enables the nucleation to start at numerous sites. In strip casting, the nucleation at numerous sites is attributed to the undercooling caused by the high cooling rate.

6.5.2 Texture in cold rolled strip
After rolling, the texture in strip cast material consists of approximately 40% deformation texture, see Figure 6.5. This texture comes close to annealed DC cast material that is rolled, which shows 33% deformation texture. As-received DC cast material consists of approximately 60% deformation texture after cold rolling due to the heavy reduction.

Comparing with literature, after 80% cold rolling the Brass component near the surface is 14% and in the centre 9%. Since the volume percentage Brass component in the centre of our rolled strip added with Si, Fe, or Cr, differs approximately 3% while the Copper component remains constant (Figure 6.11) the differences in texture between surface and centre are negligible and within the range of error. This is confirmed by [18] who reported that there is no difference between the volume percentages of Copper and Brass. This is in disagreement with [16, 17] who found a larger volume percentage Brass in the centre of the strip after cold rolling.

6.5.3 Effect of final annealing on texture in strip cast material
Final annealing results in a large decrease (-30%) of the deformation texture and a small increase (+3%) of the recrystallisation texture. It is seen that the transformation from deformation to recrystallisation texture components was in advantage of the random component which occurs due to Particle Stimulated
Nucleation (PSN). Many workers have reported that PSN as a recrystallisation mechanism will suppress the formation of Cube texture and favours a random texture [8, 18]. PSN is in favour in case of strip cast material since otherwise a too strong recrystallisation texture is obtained [19]. The strips show now volume percentages deformation and recrystallisation textures of both 10 %.

The DC cast material shows after final annealing volume percentages deformation and recrystallisation textures of both 20 %. From the microstructure it is known that the DC material is recrystallised. Here the recrystallisation mechanism was in favour of the formation of recrystallisation texture (+15 %). When the DC cast material is annealed prior to cold rolling and then final annealed, the volume percentages deformation and recrystallisation textures are 20 % and 15% respectively. In this case the recrystallisation mechanism was in favour of the formation of the random component (+14%) which is similar for strip cast material. It is seen that the texture of recrystallised strip cast material is weaker than DC material which is in agreement with [11, 19].

6.5.4 Effect of composition and treatment on the texture in strip cast material
The as-cast textures of the different compositions were all random whereas all the textures after cold rolling showed a deformation texture. In general, it is found that after rolling followed by annealing for 10 minutes the deformation textures changes marginal. Since the microstructure does not show recrystallised grains (Chapter 5), the small increase of the volume percentage recrystallisation texture can be an indication that recrystallisation just has started. Further, increasing of the volume percentage deformation texture (AlMnMg+Cr) is probably related to the coarsening of grains (Chapter 5). Since the random component decreases, it is possible that randomly oriented grains dissolve in favour of coarsening of grains that have a deformation texture. The influence of Fe on the recrystallisation texture as a result of the difference in solid solution level which is found in literature [20] could not be confirmed because no recrystallisation occurred.

In general, the inter-anneals do not change the texture except for CP Al as can be seen in Figure 6.13. The microstructure of CP Al alloy shows a recrystallised grain. The deformation texture components decrease drastically whereas the recrystallisation texture components and random component increase. The texture measurements and microstructures are in agreement. Due to the low solid solution level recrystallisation takes place [11, 21, 22].
6.5.5 Texture in strip cast material versus DC cast material

The experimentally obtained textures for strip cast material (1 mm) are presented in Figure 6.19.

![Diagram showing process route from strip cast material to rolled strip. After each step the corresponding texture is given. D: volume percentage deformation texture; RX: volume percentage recrystallisation texture.](image)

It is seen that the texture in strip cast material after cold rolling is almost equal to that in DC cast material which is annealed prior to rolling, see Figure 6.20. In addition, strip cast material does not have to start with a large volume percentage recrystallisation texture components.

In DC cast material a balanced texture is obtained after final annealing. It is seen that also for strip cast material a balanced texture is obtained after a final anneal. This texture comes close to a random texture and is weaker than DC cast material after a final anneal. Further, minimal earing is obtained when strip cast material is homogenised. The earing percentage is lower than with DC cast material.
Figure 6.20. The process route from as-received DC cast material to rolled strip. After each step the corresponding texture is given. D: volume percentage deformation texture; RX: volume percentage recrystallisation texture.

The strip cast material gives better results than the DC cast material for two reasons. First, the volume percentage texture components in the strip cast material is smaller than that of DC cast material after final annealing and therefore it will give a lower percentage earing. Second, the decrease of deformation components (Copper, R/S) into a random component in strip cast material after final annealing is larger than is seen for DC cast material. Therefore PSN is favoured in strip cast material and gives a balanced texture. As a consequence there are no compensating recrystallisation texture components needed before rolling to obtain a balanced texture.
6.6 Conclusions

Strip cast material:
- gives after casting a random texture which shows no variation throughout the thickness.
- gives after cold rolling a typical deformation texture with a lower texture strength than occurs in conventionally produced material.
- shows textures after rolling that are hardly influenced by homogenising before rolling, by rolling with inter-anneals, and by minor additions of alloying elements.

Final annealing of strip cast material that results in recrystallised strip:
- shows a balanced texture with both volume percentages deformation and recrystallisation texture components of 10%.
- shows that this is attributed to the PSN recrystallisation mechanism and results in a large random component.
- shows that the volume percentage recrystallisation texture in as-cast strip is enough to compensate the deformation texture after cold rolling, in contrast to DC cast material that needs 20-45%.
- shows a weaker balanced texture and a smaller planar anisotropy than DC cast material and gives therefore a lower percentage earing.

The formability of the strip cast material will be further improved when the casting technique and the process route are optimised.

References
Texture in strip cast material


Strip Casting of Aluminium Alloys


A The heat transfer coefficient in single-roll strip casting

A.1 Introduction

The metallurgical properties of the sheet and the productivity of the process are strongly determined by the thermal conductance at the interface between the solidifying molten metal and the cold substrate of the wheel because conductance controls phase nucleation and coarsening of the microstructure. Therefore it is one of the most important parameters in the strip casting process and is usually quantified by an interfacial heat transfer coefficient $h$, which is defined as $q = h(T_s - T_m)$, where $q$ is the heat flux and $T_s$ and $T_m$ are the surface temperatures of the strip (casting metal) and the wheel (mould) respectively. The interfacial thermal conductance is time dependent, because the contact between substrate and molten metal changes when the metal cools down and solidifies. Hence, the value of $h$ is high in the initial state of the process and decreases when an air gap in the order of microns develops. The value of $h$ is also affected by surface properties including geometric shape and dimensions of the casting, properties of the substrate material, substrate surface roughness and initial substrate temperature.

Determination of the heat transfer coefficient is possible by measuring the temperature inside the substrate at various depths with thermocouples [1, 2] and calculating the heat fluxes. In this research $h$ will be obtained by temperature measurements at various locations inside the wheel and matching these data to the calculated data with $h$ as a fitting parameter.

Literature values of $h$ are reported for different substrate materials and casting conditions. The only known values for single-roll strip casting range from 0.8 kW/m²K [2] for aluminium on a copper substrate till 20 kW/m²K [3] for stainless steel on a copper substrate, in case of a similar thickness of the strip and casting velocity. The objective of this research is to determine the heat transfer coefficient between aluminium strip (AA 3004) and copper wheel during single-roll strip casting.
A.2 Experimental

The heat transfer coefficient is obtained by recording the temperature inside the wheel during casting and this data is used in an analytical model.

![Diagram](image)

Figure A.1. Cross-section of the wheel with thermocouples located at 1 mm, 3 mm, and 6 mm beneath the contact surface. Substrate thickness: 50 mm.

A.2.1 Apparatus and material

In our laboratory we produce 1 to 3 mm thin aluminium strip with a single-roll strip caster. Liquid AlMnMg-alloy is poured with a temperature of 700 °C into a vertical delivery system positioned on top of the rotating wheel, which has initially a temperature of 20 °C. The wheel has a diameter of 0.30 m and is made of a CuCr alloy. As soon as the liquid contacts the wheel, heat is subtracted from the strip into the wheel (uni-directional freezing). The wheel carries the cooling strip away with the same velocity as the wheel and finally the strip will be peeled off after solidification of the molten metal. Since the feeding system is positioned with an angle of 20° from the center of the wheel, the maximum contact length is 110°, which is 460 mm. The temperature inside the wheel is measured by thermocouples as a function of time. To measure the temperature differences in the depth of the wheel, thermocouples are located beneath the contact surface, at depths of 1, 3 and 6 mm (Figure A.1). In this way it is possible to monitor the temperature gradients as a function of time. Signals from the thermocouples were transferred from the rotating wheel to the data system using infra-red.
A.3 Analytical model

For the analytical description, the casting process is simplified by considering a mould, initially at temperature $T_0$, with an infinite length in which a semi-infinite amount of metal, initially at its pouring temperature ($T_p$), solidifies (Figure A.2). The assumption has been made that the curvature of the wheel can be ignored because the radius of the wheel is much larger than the thickness of the strip. To solve this one-dimensional model we used an analytical method [4] that takes into account gradients within mould and casting and interface resistance between mould and casting. A virtual plane with a central temperature $T_s$ represents the temperature drop between metal and mould. This plane represents the sum of the reciprocal interface resistance, $h^{-1} = h_m^{-1} + h_s^{-1}$, where $h_m$ and $h_s$ represent the heat transfer coefficients at the wheel side and the strip side respectively. Both quantities are determined by the properties of the wheel and strip (Table A.1). During solidification heat is supplied from a source at temperature $T_s$ to the surface of the mould ($T_{sh}$) via the heat transfer coefficient $h_m$.

![Figure A.2. Schematic representation of the casting process. $T_{sh}$ and $T_{sc}$: surface temperatures of mould and casting respectively. $M$ is the solidified thickness in the x-direction; other quantities are explained in the text.](image)

The temperature profile in the mould is given as follows:
\[
\frac{T - T_0}{T_S - T_0} = \text{erfc} \left( \frac{x}{2\sqrt{\alpha t}} \right) - \gamma \text{erfc} \left( \frac{x}{2\sqrt{\alpha t}} + \frac{h_m}{k} \sqrt{\alpha t} \right)
\]

(A.3)

where $\alpha$ and $k$ are the thermal diffusivity and the thermal conductivity, respectively, and

\[
\gamma = \frac{h_m}{k} \sqrt{\alpha t} \left( \frac{x}{\sqrt{\alpha t}} + \frac{h_m}{k} \sqrt{\alpha t} \right)
\]

(A.4)

and

\[
h_m = \left[ 1 + \frac{k \rho C_p}{k' \rho' C'_p} \right] h
\]

(A.5)

where $k$, $\rho$, $C_p$ are the thermal properties of the wheel and $k'$, $\rho'$, $C_p'$ are the thermal properties of the solid metal. $T_s$ is calculated from the thermal properties of both the wheel and the solidifying metal [2, 4] which values are given in Table A.1. With $T_s$, the temperatures inside the wheel are computed as a function of time and are compared to the temperatures measured inside the wheel for the different thermocouple positions.

**Table A.1. Properties of metal and mould.**

<table>
<thead>
<tr>
<th></th>
<th>mould Cu-Cr</th>
<th>aluminium alloy: AlMnMg</th>
</tr>
</thead>
<tbody>
<tr>
<td>Specific heat $C_v$ (J kg$^{-1}$ K$^{-1}$)</td>
<td>350</td>
<td>900</td>
</tr>
<tr>
<td>Density $\rho$ (kg m$^{-3}$)</td>
<td>8890</td>
<td>2700</td>
</tr>
<tr>
<td>Conductivity $k$ (W m$^{-1}$K$^{-1}$)</td>
<td>188</td>
<td>156</td>
</tr>
<tr>
<td>Latent heat of fusion $H_f$ (J kg$^{-1}$)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Thermal diffusivity $\alpha$(*10$^5$ m$^2$/s)</td>
<td>6.04</td>
<td></td>
</tr>
</tbody>
</table>

A.4 Results

Figure A.3 represents the measured temperatures and two calculated temperature profiles obtained from the model. As the time increases the temperature rises inside the wheel and saturation is observed after about 15 seconds. Also the temperature increases with decreasing depth. The figure shows that the initial temperature rise (1 s) matches a value of $h$ of 1 kW/m$^2$K at a depth of 1 mm, and that the subsequent
The heat transfer coefficient in single-roll strip casting

temperature evolution is slower than the one, calculated with this value, indicating a trend to lower $h$ values.

![Graphs showing temperature evolution over time for different depths and heat transfer coefficients.](image)

**Figure A.3.** Measured temperatures (●) at depths of 1 mm (a), 3 mm (b), and 6 mm (c) compared with calculated temperatures for values of $h$ of 1 kW/m²K (solid line) and 0.5 kW/m²K (dotted line). The revolution time of the wheel is 1.6 s.

This observation is not clear for the other two depths but is confirmed in Figure A.4 which shows the temperature for the three positions inside the wheel as a function of time compared to calculated temperatures with an $h$ value of 0.5 kW/m²K. The calculated profiles show that in the first millimetre inside the wheel a different $h$ value applies because the experimental temperature is higher than the calculated temperature. This effect is stronger at the surface confirming a more sensitive response to the cast liquid.
Figure A.4. Temperature distribution inside the wheel as a function of time after: 0.7 s (●), 2.1 s (□), 3.5 s (△). The experimental data points are fitted with h = 0.5 kW/m²K.

A.5 Discussion

The obtained value of $h$ is low compared to several literature values. Values for $h$ in the single-roll strip casting process are reported for two different casting metals on a copper wheel, i.e. aluminium and stainless steel. Aluminium gives a value of approximately 0.8 kW/m²K [2] with a linear velocity of 0.63 m/s and a strip thickness of 1 mm, and stainless steel gives a value of 20 kW/m²K [3] with a linear velocity of 0.5 m/s and a strip thickness of 0.6 mm. Values for $h$ in the twin-roll strip casting process are reported for aluminium on a copper wheel and extend from 8.5 till 10 kW/m²K [3] with a similar strip thickness and casting velocity.

The low value of our heat transport coefficient can be explained by the fact that it is an overall value during strip casting and that there remains no optimal contact between substrate and wheel. This is because the strip comes off the wheel when the aluminium is solid, thus decreasing the overall heat transfer coefficient. This is supported by the fact that several authors [5-7] confirm a decreasing value of $h$ with decreasing contact, such as caused by increasing surface roughness or decreasing contact pressure. Apparently, formation of an air gap is promoted when solidification shrinkage is high (aluminium alloys) and when no counter force is available pressing the strip on the wheel (single-roll strip casting). The process is well described when a $h$ value of 35 kW/m²K is taken (Chapter 3).
A.6 Conclusions

The value of $h$ can be determined very well via in-situ temperature measurements inside the wheel. A simple one-dimensional model gives a good prediction of the temperatures inside the wheel and the obtained value of $h$ decreases with time (from 1.0 kW/m$^2$K to 0.5 kW/m$^2$K). The decreasing value of $h$ is attributed to the fact that heat transport into the wheel is reduced when the strip comes off the wheel as soon as it is solidified.

References

Summary

In this thesis the influence of casting conditions on the properties of single-roll strip cast strip is studied. Strip casting is an interesting casting technique because of the metallurgical and economical advantages. It is metallurgical interesting because of the rapid solidification that takes place. This can lead to improved strength and formability properties and it can lead to dedicated materials. It is economical interesting because continuous thin strip is produced and due to this thin strip hot rolling is totally eliminated. Since hot rolling is an expensive production step, strip casting is an energy and time saving process.

Many experiments were performed on a laboratory single-roll strip caster. It turned out that strip is obtained under specific conditions which makes the operation window very small. In addition, it leaves less space to study the effect of different casting parameters on the strip thickness. Therefore the process is simulated with this advantage that more insight is gained about flow in the feeding system and solidification of the liquid aluminium onto the wheel. From the experimentally obtained strip the microstructures, tensile properties and textures are studied as a function of casting conditions.

The single-roll strip cast process

From experiments it is concluded that strip of good quality is produced with a feeding system that consists of a flow modifier and an exit slit of 1.5 mm. The flow modifier contains tubes with a certain diameter and is placed just before the exit slit. Simulations showed that the flow modifier introduces some recirculations of the fluid inside these tubes. This gives the advantage that the fluid is forged to flow on that position towards the exit slit instead of flowing around inside the feeding system. The flow modifier distribute the liquid equally throughout the feeding system and that results in a constant feeding velocity of the exit slit. In addition it results in strip without gaps.

The surface quality of the strip is influenced by the flow condition. Experiments show an improved quality under constraint flow conditions and this is confirmed by simulations that do not show turbulence during the increase of the strip thickness.
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The strip thickness is influenced by many parameters. From simulations it is concluded that the thickness of the strip is mainly determined by the contact time between wheel and liquid.

Properties of aluminium strip cast material

Optical microscopy of various strip cast AlMnMg alloys shows that the grains are equiaxed throughout the whole strip in absence of columnar crystals. This is in contrast to the solidification theory about unidirectional solidification. The strip solidifies with a cooling rate of 800 K/s which is for conventionally cast aluminium in the order of 100 K/s. In contrast, a relatively large average grain size (140 μm) is found. Remelting of the growing grains cause the relatively large grain size for the strip cast material.

Rolled strip cast material shows elongated grains and aligned intermetallics. This is similar to conventional cast material. The size of the intermetallics depend largely on alloy composition whereas minor additions of Si, Fe, or Cr to the AlMnMg alloy hinder growth of intermetallics during rolling. Precipitation from the saturated solid solution at room temperature occurs if the accumulated strain is sufficiently high. Recrystallisation temperature and time is increased for strip cast material due to a larger saturation level which is attributed to the higher cooling rates. The mechanical properties of rolled strip cast material are within the property limits of the ASTM standard.

Strip casting gives a random texture and the directional dendrites found in the microstructure causes no texture variation throughout the thickness. Cold rolling gives a typical deformation texture with a lower texture strength than occurs in conventionally produced material. Homogenising before rolling, rolling with inter-anneals, and minor additions of alloying elements have no influence on the strength of the deformation texture.

Final annealing of strip cast material that results in recrystallised strip shows a balanced texture with both volume percentages deformation and recrystallisation texture components of 10 %. The recrystallisation is attributed to the Particle Stimulated Nucleation mechanism and results in a decrease of the deformation texture component and a large random component. Further, it turned out that the volume percentage recrystallisation texture in as-cast strip is enough to compensate the deformation texture after cold rolling, in contrast to DC cast material that needs 20-45 %.
The balanced texture that is obtained in strip cast material after a final heat treatment is weaker than in DC cast material. In addition, strip cast material has a smaller planar anisotropy in that condition and gives therefore a lower percentage earing than DC cast material. This gives improved deep drawing properties.
Samenvatting

In dit proefschrift is de invloed van gietcondities op de materiaaleigenschappen van direct dun gegoten aluminium strip onderzocht. Het direct dun gieten van strip op een gekoeld ronddraaiend wiel, is een interessante gietmethode omdat het metallurgische en economische voordelen oplevert. Het gekoelde wiel zorgt voor een snelle stolling van het vloeibare aluminium. Snelle stolling geeft een verfijnde microstructuur in combinatie met een oververzadiging van elementen in oplossing wat kan resulteren in betere mechanische en vervormingseigenschappen (in vergelijking met conventionele giettechnieken) en nieuwe materialen. Het is een economisch interessant proces omdat het direct dunne strip geeft en daarom geen warmwalsbehandeling nodig heeft. Dit scheelt aanzienlijk in de kosten vanwege tijd en energie-besparingen.


Het dunne stripgieten

Uit experimenteel onderzoek blijkt dat 1.4 mm dunne strip met een goede kwaliteit wordt verkregen met een toevoersysteem dat is uitgerust met een zogenaamd stromingsverfijner en met een uitstroomspleet van 1.5 mm. Deze stromingsverfijner bestaat uit een aantal cilindervormige holtes en zit net boven de uitstroomspleet. Simulaties toonden aan dat deze holtes wervelingen veroorzaakten waardoor de vloeistof geforceerd werd om in de holtes te stromen. Aangezien dit in elke holte plaatsvindt, stroomt de vloeibare aluminium gelijkmatig naar de uitstroomspleet zodat een constante voeding ontstaat. Vervolgens resulteert dit in een strip van gelijkmatige dikte en zonder gaten.
De oppervlaktekwaliteit van de strip wordt beïnvloed door de manier van uitstromen. Een geforceerde stroming, dat betekent dat de vloeistofstroom tussen
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wiel en toevoersysteem ingeklemd stroomt en vervolgens stolt, geeft betere resultaten dan vrije stroming. Dit wordt bevestigd door de simulaties.

De stripdikte wordt beïnvloed door vele parameters. Simulaties toonden aan dat de stripdikte voornamelijk wordt bepaald door de contacttijd van het aluminium met het wiel.

Eigenschappen van aluminium stripgiets materiaal

Uit optische microscopie van een aantal stripgegoten AlMnMg legeringen blijkt dat door de gehele dikte van de strip de korrels equiaxiaal zijn en dat er geen kolomstructuur is. Dit wijkt af van de stollingstheorie over uni-directionele stolling maar daarentegen is de afwezigheid van de kolomstructuur gunstig voor de vervormingseigenschappen. De afwezigheid kan verklaard worden door de snelle stolling en het vervolgens direct ontstaan van een luchtgat, in de orde grootte van microns, tussen wiel en strip. De stolsnelheid van de strip is 800 K/s (conventioneel gieten zoals Direct Chill gieten, levert een stolsnelheid van 100 K/s op). De korrelgrootte daarentegen is vrij groot, 140 μm, maar dit komt door het hersmelten van groeiende korrels.

Gewalste strip laat lange korrels en intermetallische verbindingen zien die in de walsrichting liggen. Dit vertoont hetzelfde beeld als conventioneel gegoten en gewalst material. De afmetingen van de intermetallische verbindingen hangen af van de legeringssamenstelling, alhoewel kleine toevoegingen van Si, Fe, of Cr aan een AlMnMg legering de groei tijdens het walsen hinderen. De rekristallisatietermperatuur en tijd van aluminium stripgiets materiaal worden verhoogd door het stripgieten. De snelle stolling zorgt voor meer elementen in oplossing waardoor de rekristallisatie wordt bemoeilijkt. Gehomogeniseerd stripgiets materiaal verlaagt de rekristallisatietemperatuur doordat meer elementen uit de oplossing zijn. De mechanische eigenschappen van gewalst aluminium stripgiets materiaal voldoen aan de ASTM norm.

Stripgieten geeft een willekeurige textuur (oriëntatie van de korrels) door de gehele dikte van de strip. Kouwdansen geeft een typische deformatie textuur met een intensiteit die lager is dan voor conventioneel geproduceerd materiaal. Homogeniseren voor het walsen, walsen met tussentijds gloeien, en kleine toevoegingen van legeringselementen, hebben geen invloed op de intensiteit van deze deformatie textuur. In koudgewalst stripgiets materiaal, wat vervolgens een warmtebehandeling heeft gekregen, worden gelijke intensiteiten deformatie-
textuurcomponenten en rekristallisatie-textuurcomponenten gevonden. Het merendeel van de textuur is willekeurig. Dit wordt veroorzaakt door het optredende rekristallisatie mechanisme, het zogenaamde Particle Stimulated Nucleation mechanisme. Dit mechanisme vindt juist plaats bij stripgietmateriaal vanwege de vele kleine intermetallische fasen. Het mechanisme zorgt ervoor dat korrels met een deformatietextuur zodanig roteren dat er willekeurige orientaties ontstaan. Dit mechanisme is voor stripgietmateriaal gunstig, in tegenstelling tot conventioneel materiaal, aangezien het volume percentage van de overblijvende deformatie-textuurcomponenten gelijk wordt aan de rekristallisatie-textuurcomponenten. De vervormbaarheid van het materiaal neemt toe wanneer de volume percentages van de twee typen componenten gelijk zijn (=gebalanceerd) en laag zijn. De gebalanceerde textuur in stripgietmateriaal na de laatste warmtebehandeling is zwakker dan in Direct Chill materiaal. Daardoor heeft stripgietmateriaal een kleinere planaire anisotropie en een lager percentage oorvorming in vergelijking tot Direct Chill gietmateriaal. Dit geeft dus verbeterde dieprekeigenschappen. Optimalisatie van de warmtebehandelingen voor stripgietmateriaal zal de vervormbaarheid verbeteren.
Dankwoord

Bij deze wil ik mijn promotor Laurens Katgerman bedanken voor de enorme leerzame en leuke tijd die ik dankzij hem heb mogen ervaren. Ik heb ontzettend veel geleerd in mijn promotie tijd wat mogelijk was door Laurens manier van begeleiden die ik als zeer prettig heb ervaren.

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Strip Casting of Aluminium Alloys

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List of Publications

Journal papers:


Conference papers:


Curriculum Vitae

Etnel Straatsma

geboren op 22 februari 1971 te Wolvega

1983-1988 HAVO, Andreas Scholengemeenschap Zevenaar
1988-1990 VWO, Liemers College Zevenaar
1990-1991 Grafische Vormgeving, Hogeschool van de Kunsten te Utrecht
2001 Manufacturing engineer eq. product technologisch onderzoeker op het gebied van plaatomvormen bij de unit Onderdelenfabricage van Stork Fokker te Papendrecht.