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Factors affecting thermal contraction behavior of an AA7050 alloy

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ABSTRACT

The understanding of the contraction behavior in high-strength wrought 7XXX-series alloys is very important for the analysis of stress–strain development and modeling of hot tearing and cold cracking in direct-chill casting. The linear thermal contraction during and after solidification of an AA7050 alloy has been studied experimentally under different conditions. The experimental parameters included the sample height (metal level in the mold), grain refining, and gas content. The results showed that evolution of gas during solidification and pressure drop across the mushy zone are the main reasons for the preshrinkage expansion. The measured coefficient of linear thermal contraction decreases with the decreasing grain size at all temperature ranges. The role of gas precipitation and grain refining is discussed.

1. Introduction

Direct-chill (DC) casting is the major technology for producing aluminum ingots and billets for further deformation processing. The products have been widely applied in the automotive, aerospace and construction industries. However, casting defects arising during and after solidification of metallic alloys constraint their applications.

Among the most common problems that occur during casting are hot tears and cold cracks. The main cause of these defects is that stresses and strains built up during and after solidification are too high compared to the actual strength and ductility of the semisolid and as-cast material [1,2]. Cracking is particularly typical of DC casting of high-strength aluminum alloys, such as an AA7050 alloy.

Generally, an AA7050 alloy appears to be extremely brittle in the as-cast condition at temperatures below 300 °C [3]. Large solidification range and low thermal conductivity values [4] make the DC casting billets from this type of alloys vulnerable to both hot and cold cracking due to imposed large thermal gradients. Non-equilibrium solidification conditions, on the other hand, result in micro-segregation and formation of non-equilibrium phases (intermetallics) mainly on the grain boundaries and inter-dendritic space. Such brittle phases potentially provide favorable locations for crack initiation and propagation. Despite considerable efforts that have been made to avoid the hot and cold cracking, this kind of defect is still a considerable problem in casting of high-strength aluminum alloys.

In the manufacturing practice, grain refining, filtration and degassing are the common ways to alleviate the hot and cold cracking [1]. During casting, grain refiner is added and degassing is applied to achieve fine equiaxed grains and reduce the porosity, which can decrease the sensitivity to cracking and improve the mechanical properties of the as-cast metal [2,5,6]. But the specific effect of these methods on reducing the driving forces for cracking is still unclear and few reviews are available on this subject. As for the driving force, it is generally accepted that the inadequate feeding compensation of the shrinkage and solid contraction in the presence of thermal stresses caused by temperature gradients in the solidifying casting is the major reason for the occurrence of hot cracking [1,7]. As for the cold cracks, the accumulation and concentration of residual stresses in the casting during cooling after the end of solidification under conditions of high temperature gradients and low ductility is usually given as a main reason for the brittle failure [8]. Thermal contraction above and below the solidus is, therefore, the constitutive property that governs the formation of stresses and eventually the cracking of the casting. The understanding of the contraction behavior is very important for the analysis of stress–strain development and modeling of hot tearing and cold cracking.

It was noticed that the apparent thermal contraction of the billet shell surrounding the liquid and semi-liquid sump during DC casting was larger than it could be estimated based on the coefficient of

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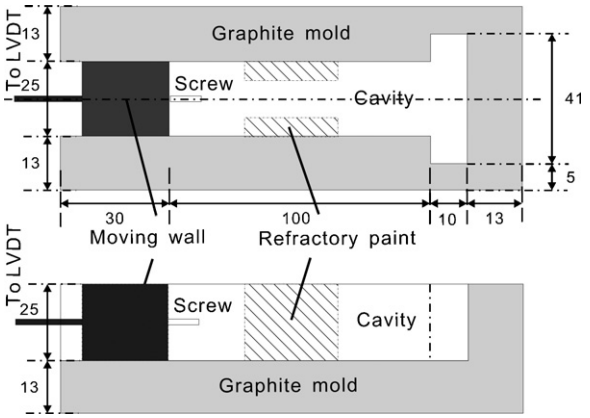

Fig. 1. Schematic view of the experimental setup. Dimensions are given in millimeters.

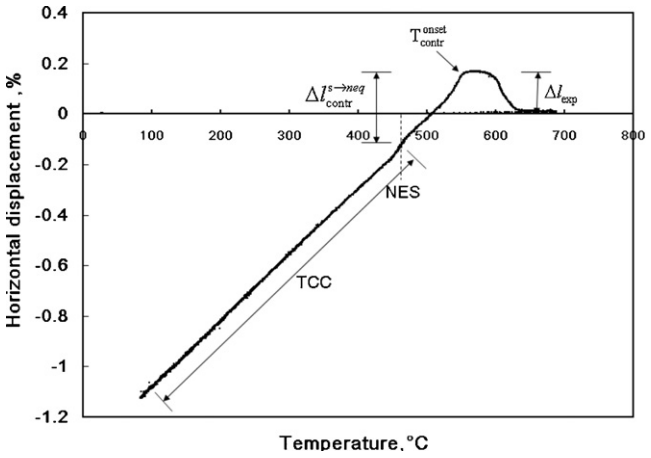

Fig. 2. Examples of data obtained from experiments. Δl_{exp} is the preshrinkage expansion; T^{onset}_{contr} is the temperature of the contraction onset; Δl^{s-neq}_{contr} is the amount of contraction in the non-equilibrium solidification (NES) range, and TCC is range where the linear thermal contraction coefficient (TCC) is calculated.

thermal expansion [9]. This happens because the internal volume of the billet, being semi-liquid and soft, offers little resistance to the thermal contraction of the external shell and the rate of contraction becomes higher as compared to the completely solid billet. The "additional" thermal contraction increases when the billet sump becomes deeper, i.e. with the increased casting speed and billet diameter [9]. These early observations drew attention to the relationship between the inhomogeneity of solidification development and the thermo-mechanical behavior of the real casting.

In the past decade, an experimental technique for measuring the various contraction parameters has been developed and employed [10]. It was shown that the geometry of the sample and structure of the alloy affect the contraction behavior in the solidification range. It was also shown that the same technique can be successfully used for estimating the linear thermal contraction coefficient at high temperatures. This paper presents the results on the effect of such process parameters as grain refining and gas saturation on the contraction behavior of an AA7050 alloy at temperatures above and below the solidus.

2. Experimental procedure

The experimental setup for measuring the linear contraction at temperatures above and below the solidus is described in detail elsewhere [10]. It is based on the idea suggested by Novikov [6] and comprises the following parts: a T-shape graphite mold with one moving wall, a water-cooled bronze base which provides high cooling rate that is similar to the conditions in DC casting, a linear displacement sensor (linear variable differential transformer (LVDT)), a K-thermocouple, and a computer-based data acquisition system.

Fig. 1 shows the schematic view of the experimental setup. The cross-section of the main cavity is 25 mm × 25 mm with a gage length of 100 mm. The cross-section of the T-shape cavity is thinner than that of the main cavity, which allows the melt to solidify faster. So the solidifying sample can be fixed on this side. On the opposite side, a metallic rod is fixed in the moving wall to attach the solidifying metal, so the position of the moving wall can be detected by LVDT, which is accurate to 6 μm or 0.006% of the gage length. The temperature is monitored with a 0.1-mm thin, open-tip K-thermocouple standing vertically in the center of the mold at about 1.5 mm above the mold bottom to avoid the problem of filling the gap between the thermocouple tip and the mold bottom. During the experiments, the temperature and displacement are recorded simultaneously by the data acquisition system.

Based on the previous experience [10], a refractory paint (bone ash) was applied onto the internal walls in the central part of the mold, as shown in Fig. 1, to equalize cooling rates, reduce temperature gradient, and assure that the first contact between opposite coherent parts of the sample occurs in the area where the temperature is measured. The melt temperature was 720 °C in all studied cases.

After having acquired the primary data, the temperature was plotted versus time in order to derive the cooling curves from which the cooling rate and characteristic temperatures in the solidification range were obtained. After this, the data were reconstructed to find the direct dependence of displacement on temperature. Fig. 2 demonstrates an example of the contraction curve for an AA7050 alloy. From this curve, the temperature of the contraction onset, the preshrinkage expansion, the amount of contraction in the solidification range and the thermal contraction coefficient (TCC) at subsolidus temperature can be extracted directly.

The preshrinkage expansion is calculated in the temperature range between the temperature of expansion onset and the temperature of contraction onset.

The linear solidification contraction (the amount of contraction in the solidification range) is determined as follows [6,10]:

$$\varepsilon_s = \frac{l_s + \Delta l_{exp} - l_f}{l_s} \times 100\%, \quad (1)$$

where l_s is the initial length of the cavity (100 mm); l_f is the length of the sample at the temperature of non-equilibrium (NEQ) solidus, and Δl_{exp} is the preshrinkage expansion. It is worthy to note that at the used cooling rates (5–10 K/s), the solidification range was considered as NEQ, hence the solidification ended at the eutectic temperature. The temperature of non-equilibrium solidus (NES) for the AA7050 alloy is about 465 °C, which is observed from the cooling curves and agrees well with literature data and Thermocalc calculations [11].

The average TCC is calculated as follows:

$$TCC = \frac{L_{T_2} - L_{T_1}}{L_{gage}(T_2 - T_1)}, \quad (2)$$

where T_2 and T_1 are the temperatures ($T_2 > T_1$) below the solidus; L_{T_2} and L_{T_1} are the positions of the displacement sensor at T_2 and T_1, respectively; and L_{gage} is the gage length of the sample. Experiments were performed with an AA7050 alloy, which composition is given in Table 1. The melt was prepared in the electrical resistance furnace in graphite crucibles. When studying the effect of grain refinement different amounts of Ti (from 0.01 to 0.2 wt%) in a form of an Al–8% Ti master alloy were added to the melt at 740 °C.