

The Effect of Pouring and Vibration on Cast Quality.

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ABSTRACT

The effect of vibration and pouring on the texture and mechanical properties of Aluminum-Copper alloys, centerlines, shrinkage, and segregation in casting of metals had an adverse effect on cast product quality. Nozzle electromagnetic stirring (N-EMS) is an effective method to refine the solidification structure of casting of metals thereby improving it. The effect of N-EMS on the solidification structure of continuous cast alloy has been investigated by means of experiments. The results show that, upon imposing N-EMS, crystal grains within casted materials are refined and continuous casting can be conducted at a lower pouring temperature; the solidification structure of the resulting structure is much finer than without N-EMS. The mechanism by which the solidification structure is improved is considered to be a change in the flow and temperature gradient of molten metal.

Three compositions of alloys were made for each sample; five specimens were vibrated at different frequencies during solidification (cast). The results showed that the grain size decreases (fine grained structure) with frequency. The hardness, strength increased, and ductility decreased. Gas and shrinkage porosity were also reduced.

(Keywords: vibration, solidification, pouring techniques, hardness, strength, ductility)

INTRODUCTION

In casting, a solid is melted through heating to a proper temperature (melting temperature), the molten material is then poured into a cavity or mould which contains a replica (shape) and allowed to cool and solidify. The resulting product can virtually have any configuration the designer desires. In addition the resistance to working stresses can be optimized through the control of

certain properties leading to a good aesthetic appearance of the cast product [12].

Mould conditions, pouring rate, and other process variables are factors that would definitely have an effect on the microstructure and properties of the cast, [3] bearing in mind the requirement that the material solidify in a manner that would maximize the properties desired while simultaneously preventing potentials defects such as shrinkage, voids, gas porosity, and trapped inclusions. The most efficient method of early fault detection in a metal cast is high-frequency vibration analysis as its parameter changes most quickly in the early stage of the defect development. Solidification behavior is heavily influenced by the thermal gradient in the liquid and the velocity with which it moves through the casting [17].

Small scale casting techniques have been developed to minimize the need for casting of full size ingots during direct-chill casting. These techniques either mimic solidification conditions within the ingot (Dissimulator) or allow solidification to occur under controlled conditions such that the fundamentals of solidification can be investigated.

MODELS OF INITIATION OF ALPHA-AL GRAINS

The formation of stable α -Al crystallites was, until recently, almost always modeled by classical heterogeneous nucleation theory whereby a transfer of atoms from the melt to the α -Al phase is facilitated by inoculant particles [1]. In this model, the α -Al phase maintains a contact or wetting angle θ with the substrate, determined by the balance of interfacial energies σ between the three phases (denoted by subscripts l, s, and c for liquid, substrate and crystallite, respectively).

The wetting angle is given by the following:

$$\cos \theta = \frac{\sigma_{ls} - \sigma_{cs}}{\sigma_{lc}} \quad (1)$$

The work required to overcome the barrier to heterogeneous nucleation, W_{hetero}^* is given by:

$$W_{\text{hetero}}^* = \frac{16 \pi \sigma_{lc}^3}{3 \Delta G v^2} f(\theta) = W_{\text{hetero}}^* f(\theta) \quad (2)$$

where W_{homo}^* is the work required to overcome the barrier to homogenous nucleation, ΔGv is the volumetric Gibbs energy of fusion, and $f(\theta)$ is a function of the wetting angle.

$$f(\theta) = \frac{2 - 3\cos \theta + \cos^3 \theta}{4} \quad (3)$$

The rate at which potential nuclei overcome this energy barrier I (which can be expressed either per unit area of substrate surface or per unit volume of melt) is given by:

$$I = I_0 \exp \left[W_{\text{hetero}}^* + \frac{\Delta G_d}{K_B T} \right] \quad (4)$$

where I_0 = pre-exponential factor; and G_d = energy barrier to atomic additions at the crystallite – liquid interface.

When expressed per unit volume of melt, I_0 is determined by the atomic vibration frequency and the number of atoms per unit volume and implicitly by the total area of substrate per unit volume [9][13]. As θ decreases, the work required to create a stable crystallite decreases and the nucleation rate increases, the latter being highly dependent on both W_{hetero}^* and θ . However, the value of θ is rarely known accurately; consequently, it is frequently used as an adjustable parameter, as if the under cooling is sufficient for the crystallite to reach a hemispherical shape, subsequent growth will decrease the curvature of the interface; no further increase in the under cooling is required for growth to continue.[16] The undercooling ΔT_{fg} required to achieve this state of free growth is given by:

$$\Delta T_{fg} = \frac{4 \sigma_{lc}}{\Delta S v d \text{ part}} \quad (5)$$

Where d part is the diameter of the {0001} face of the inoculants particle and Δsv is the volumetric entropy of fusion. The free-growth undercooling is

inversely related to particle size; hence, as the undercooling increases, the larger particles become active first. If the free-growth undercooling is larger than that required for nucleation, the barrier to free growth determines the activation of inoculants particles. Hence, the term “grain initiation” was coined to describe the complete process of formation of a freely growing α -Al grain [15].

NATURE OF THE DRIVING FORCE (THERMODYNAMIC ASPECTS)

Considering an heterogeneous equilibrium (equilibrium involving more than one phase), the free energy G of a component phase is defined by:

$$G = H - Ts \quad (6)$$

where H = enthalpy; T = Absolute temperature; and S = Entropy.

If we consider pressure to be constant as with most metallurgical systems, then we have:

$$\frac{\Delta G}{\Delta T} = \text{Const} = -S \quad (7)$$

Thus the free energy decreases with increasing temperature.

TYPES OF NUCLEATION

There are essentially two types of nucleation and these are:

- (i) Homogeneous nucleation
- (ii) Heterogeneous nucleation.

Homogeneous Nucleation

When a solid form within its own melts without aid of foreign materials, it is said to nucleate homogeneously [14]. The energy balance on which nucleation depends has the free energy change for a spherical particle as follows:

$$\Delta G = \frac{4\pi r^3}{\gamma} - \frac{4}{3} \pi r^3 Gv \quad (8)$$

where ΔG = Free energy; $4 \pi r^{3\gamma}$ = Interfacial free energy term; r = radius of duster; γ = specific interfacial free energy per unit area, and G_v = Volume energy change.

But,

$$G_v = \frac{\Delta H \Delta T}{TE} = \text{Change in volume free energy per unit volume resulting from the formation of solid nucleus.}$$

where ΔH = enthalpy change (latent heat of fusion) and ΔT = extent of under cooling.

If $G_s = 4 \pi r^2 \gamma$, we can simplify the total free energy ΔG as follows:

$$\Delta G = -4/3 \pi r^3 G_v + G_s \quad (9)$$

Figure 1 shows the variation of ΔG with radius of the solid nucleus.

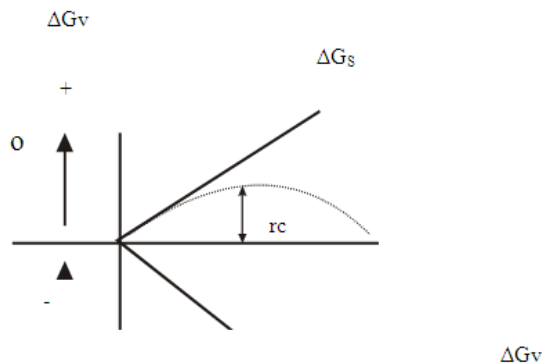


Figure 1: Variation of ΔG with Radius of Solid Nucleus.

Figure 1 also shows that at a critical radius r_c , G is maximum and here ΔG is zero and therefore we have:

$$0 = -4 \pi r^2 \frac{\Delta H \Delta T}{TE} + 8 \pi r \gamma$$

$$\text{This implies } n = \frac{2\gamma TE}{\Delta H \Delta T} \quad (10)$$

Hence the tendency for growth is to decrease ΔG , solid duster should therefore be greater than r_c for

a nucleus to grow and hence a decrease in ΔG will facilitate growth since r will increase further.

Thus particles smaller than r_c will dissolve or evaporated if the particle is a liquid in a super saturated vapor, because only in this way can the particle achieve a reduction in its free energy.

HETEROGENEOUS NUCLEATION

Here the initial interface for nucleation is produced by foreign particle present in the melt or intentionally added surface of the container (mould) or the surface of the liquid or a solid film such as oxide also produce initial interface for nucleation.

A non-spherical nucleus will have a critical volume than a spherical one and since surface area increases as a result of departure from spherical forms, and this also increases the surface energy G_s , it then follows that nucleation pattern in many liquid melt will be heterogeneous [10][11].

Volume of a Spherical Cap: Spherical cap of a solid formed on a planar substrate formed on a planar substrate. Consider Figure 2a:

If V = volume of the sphere

V_c = Critical volume = Volume of the non spherical nucleus.

$$V = 4/3 \pi r^3$$

$$\frac{V_c}{V} = \frac{1/3 \pi r^2 (3r - h)}{4/3 \pi r^3} = \frac{h^2 (3r - h)}{4r^3}$$

For $h/r \leq 2$ we have $V_c \leq V$

But in the foregoing discussion, the non-spherical nucleus usually has a very large radius. Thus, for heterogeneous nucleation:

$$h/r > 2 \Rightarrow V_c > v.$$

For contact angle θ

$$\text{Volume} = \frac{1}{3} \pi h^2 (3r - h) = \frac{1}{3} \pi r^3 (2 - 3 \cos \theta + \cos^3 \theta)$$

$$\text{Surface area} = 2\pi rh$$

$$= 2\pi r^2 (1 - \cos \theta)$$

Under stable conditions:

$$\cos \theta = \frac{n(\gamma_{sc} - \gamma_{sl})}{\gamma_{lc}} \quad (11)$$

where n is the roughness of substrate surface (i.e. mould surface).

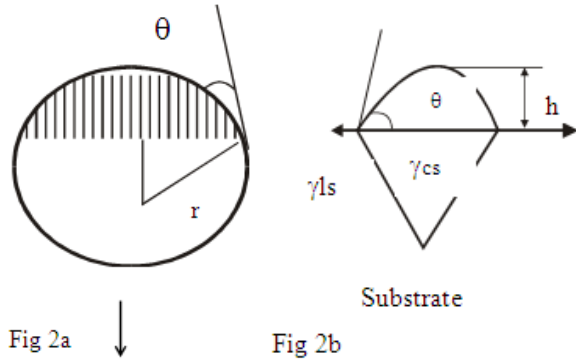


Figure 2: Volume of a Spherical Cap.

NUCLEATING AGENTS

One of the most important technical aspects of casting is the control of nucleation. Grain size and grain shape together with segregation effects on the physical and mechanical properties of the cast product. Control is exercised by the use of nucleating agents (i.e., inoculants). These inoculants that act as catalyst may react with the liquid or may be insoluble in it, and in both cases produces a small contact angle [2].

Both surface energy relationship and shape have been proposed as the two factors controlling the effectiveness of the nucleants. The surface energy relationship suggests that the substrate surface roughness decreases the angle of contact, which increases the effectiveness of the nucleants. From the equation obtained in Equation 11 for the cosine of the contact angle i.e.,

$$\cos \theta = \frac{(\gamma_{sc} - \gamma_{sl})}{\gamma_{lc}}$$

We have:

- i) if $n < 1$ (smooth surface). This implies $\cos \theta$ decreases hence θ decreases.

- ii) if $n > 1$ (rough surface). This implies $\cos \theta$ increases hence θ decreases. This means that roughness decreases angle of contact and hence increases potency of the nucleants.

PRACTICAL IMPORTANCE OF CONTROLLED NUCLEATION

Our past analysis of nucleation has been on static situation. Only temperature and the potency of nucleating substrates have been considered as having any effect. Subjecting the liquid metal to dynamic stimuli and pouring can still induce nucleation [7][13]. This can be classified into two distinct forms:

- (i) An initial nucleation phenomenon induced mechanically and
- (ii) Increasing greatly the number of crystals in a solidifying liquid by a dynamic means.

The forms of disturbance, which could really affect the method control, are friction, vibration and pressure pulse due to pouring. Since mechanical properties and microstructure are related, it follows that nucleation controls not only determines the microstructure but also brings about variation in the mechanical properties. Thus alloyed metals, improve in mechanical properties (nucleation control) such properties can be improved further by vibrating the mould during solidification.

EXPERIMENTAL PROCEDURE

Three compositions of alloys were made for each composition of Aluminum-Copper alloys of which five specimens were vibrated at different frequencies during solidification (Casting).

Both the Aluminum ingots and copper were cut into small pieces. Five sets of three different compositions were prepared. They are the 2%, 4% and 8% Cu-aluminum alloys.

Both the graphite crucible and permanent mould to be used were thoroughly clean the vibration was tested by connecting it to the main supply and then adjusting the knob to different frequencies so that a working condition was obtained [5].

The furnace (pyrometer) was put on, and the mould pre-heated to a temperature of about 200°C. The mould was now removed from the furnace and bolted on to the vibrator. Both the copper and aluminum of known composition as earlier obtained were added to the crucible which was then placed inside the furnace until the temperature was just above 1000°C as measured by a pyrometer. The furnace temperature was then slightly reduced until its temperature falls just below 760°C which is a desired pouring temperature of the alloy. At this point the crucible was brought out of the furnace after making sure that the furnace temperature has been reduced. It was placed on the floor and quickly stirred to obtain uniformity. After which the molten metal was poured into the mould, which has earlier been set into vibration to the required frequency.

After 3 minutes, the vibration was switched off and the cast metals (alloy) were noted and polished. After the final girding, each specimen was then polished on a selvyte cloth impregnated

with aluminum paste held in a liquid suspension [4]. The final surface is mirror like. This surface was highly deformed. It was then removed by etching. The purpose of etching is to make visible the many structure characteristics of the metal.

The required etching was prepared and it was made up of the following:

Hydrofluoric acid (40%)	1.0ml
Nitric acid (conc.)	2.5ml
Hydrochloric acid (conc.)	1.5ml
Distilled water	95.00ml

This etchment is good for revealing microstructure quality and grain size, particularly in aluminum alloys containing copper. The etching time was about 20sec. after which it was left under the tap for some few seconds. The specimens were vigorously shaken after removing from the running tap. They were then left in an incubator for about 40 to 60 seconds to dry.

Table 1: Variation of Percentage Reduction in Area with Frequency.

% Reduction in Area.			
Alloy Composition Frequency of Vibration (Hz)	92/8 Al – Cu	96/4 Al – Cu	92/2 Al – Cu
0	10.0	13.0	26.0
10	7.0	9.0	20.0
20	5.0	6.0	15.0
35	3.0	5.0	8.0
50	2.0	4.0	5.0

Table 2: Variation of 0.1% Proof Stress with Frequency.

0.1% Reduction in Area			
Alloy Composition Frequency of Vibration (Hz)	Al - Cu	Al – Cu	Al - Cu
0	22.21	22.20	14.10
10	40.10	35.75	21.35
20	55.34	53.21	34.73
35	58.16	56.72	37.33
50	58.70	58.01	38.45

Table 3: Variation of Grain Size with Frequency.

Grain Size Diameter (Mm) % Reduction in Area.

Alloy Composition Frequency of Vibration (Hz)	92/8 Al - Cu	96/4 Al - Cu	92/2 Al - Cu
0	73.52	67.85	63.24
10	68.37	56.27	52.18
20	60.25	48.25	47.61
35	47.15	38.25	38.23
50	43.21	34.26	33.10

Table 4: Effect of Vibration on the Fracture Stress of the Alloy.

Fracture Stress MN/M²% Reduction in Area.

Alloy Composition Frequency of Vibration (Hz)	Al - Cu	Al - Cu	Al - Cu
0	23.21	22.28	14.44
10	42.30	36.11	22.33
20	59.10	53.35	32.06
35	55.04	59.32	41.27
50	45.28	47.10	37.07

OBSERVATIONS

Ductility: An almost negative linear relationship was obtained when the percentage elongation was plotted against frequency of vibration is a measure ductility. It shows that ductility decreases with frequency.

Also from the results obtained when the percentage reduction in area was plotted against frequency. Figure 3 shows that ductility decreases with frequency. Maximum ductility was obtained at 0Hz for the three alloys while the least ductility was obtained at 50Hz. Generally the 2% cu-al-alloy composition show more ductility while the 8% cu-al-alloy metal was least ductile.

The Proof Stress: From the plotted result (Figure 4), it was observed that the proof stress (0.1% proof stress) increases with frequency. The 8% Cu-Al-metal alloy has the highest proof stress for each of the least.

The Fracture Stress: For each frequency, the 8% Cu-Al metal alloy has maximum value and the fracture stress increase with frequency up to 35Hz, after which it falls (Figure 5).

Microstructure: From the graph of grain size plotted against frequency of vibration (Figure 6). It can be observed that as the frequency increases, smaller grains formed, "giving fine texture". This was expected since the hardness and the proof stress increase with frequency while ductility falls an indication of smaller grain size to numerous grains.

DISCUSSION OF RESULTS

Figure 3 shows the variations of percentage reduction in area with frequency while Figure 4 shows the proof stress variation with frequency. It could be observed that in both cases there is an initial high increment in value, which diminishes as the frequency, increases. As can be seen, the variation of both parameters from 35Hz to 50Hz is very small especially for the proof stress. This gives an indication that the effect of variation at high frequency is very low for the range of frequency considered. From the plot of percentage elongation with frequency there shows an almost linear relationship with negative slope. The percentage reduction in area which although is a curve follows the same pattern as the % elongation.

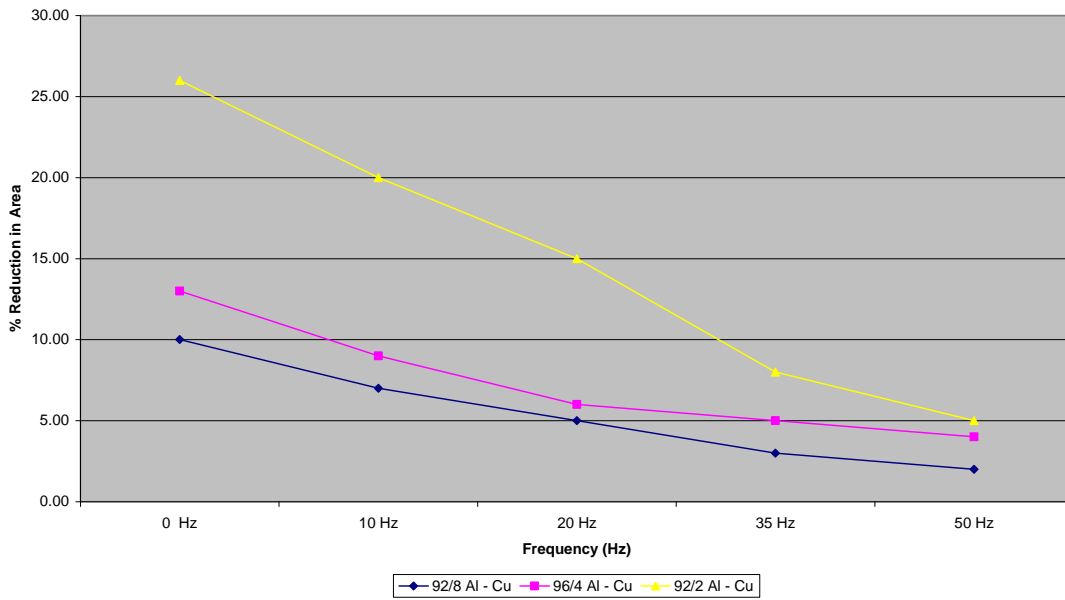


Figure 3: Variation of Percentage Reduction in Area with Frequency.

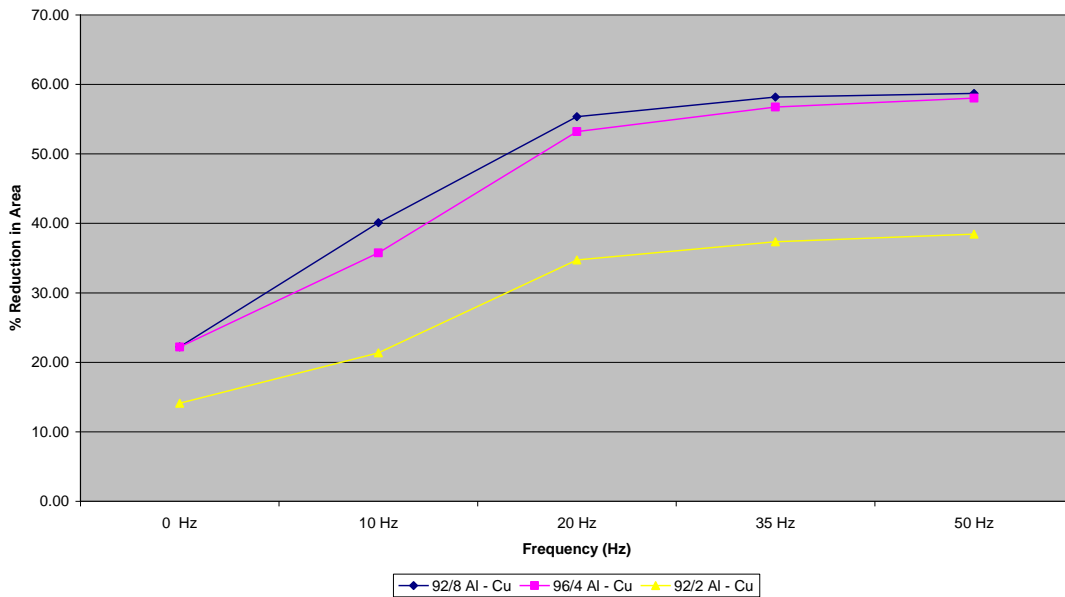


Figure 4: Variation of 0.1% Proof Stress with Frequency.

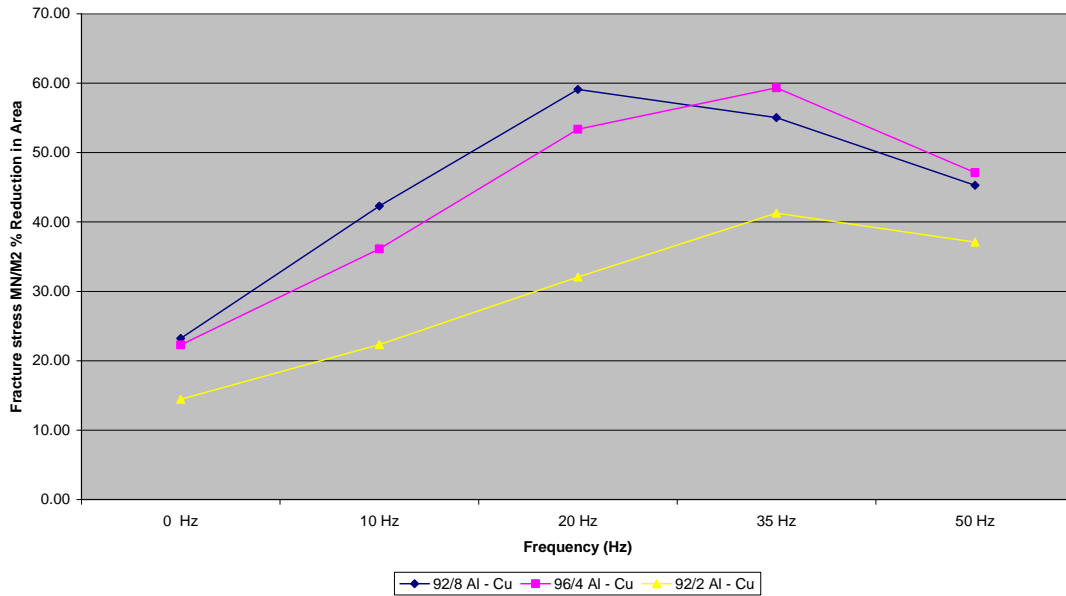


Figure 5: Effect of Vibration on the Fracture Stress of the Alloy.

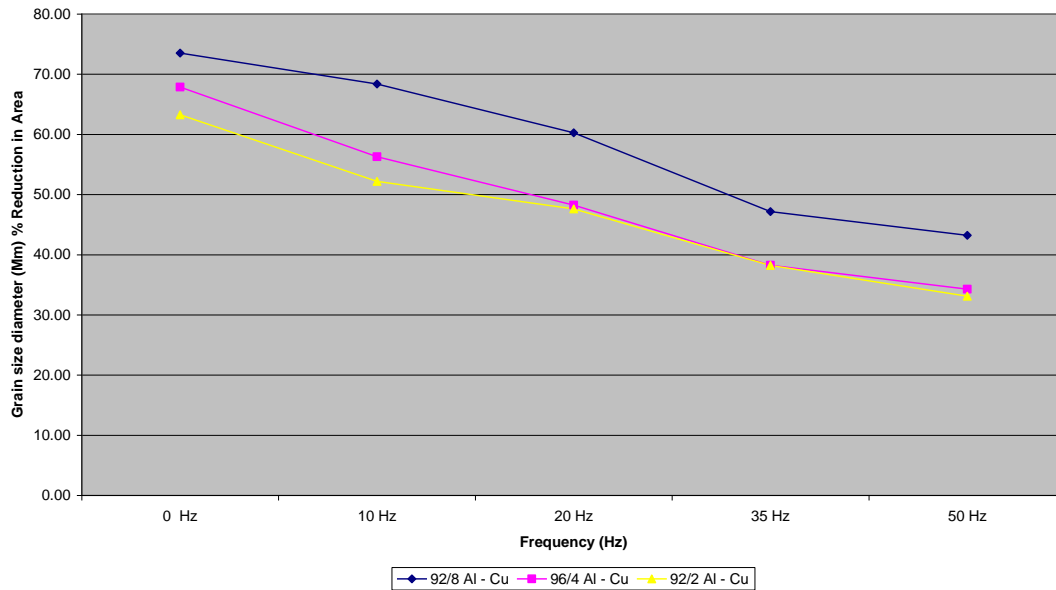


Figure 6: Variation of Grain Size with Frequency.

There is a relationship between hardness and tensile strength, and it states a direct proportionality. If the tensile stress is substituted for the proof stress (even though proof stress is usually substituted for yield stress), a direct statement is expected.

Comparing the hardness and ductility also, it was noted that a ductile material would be less capable of resisting indentation than a less ductile material. Hence, a negative relationship is expected and the results obtained confirmed this statement.

From the result obtained in the case of fracture stress against frequency of vibration, it was observed that there was an increase in fracture stress up to 35Hz and then there was a fall; although, it was expected to follow the same pattern as the hardness and “grain size” relationship. A look at the % reduction shows that at 50Hz, the value is below “5” which is that the brittle range. It is expected of a ductile material with the same strength, less ductile material to withstand more loads. This accounts for the fall in fracture strength even though a finer grain structure was obtained here.

CONCLUSION

It is observed that production of pressure tight castings from heavy alloys often results in problems reasoned by the long freezing range and as a consequence of the tendency for micro-shrinkage. Exemplified on a copper casting alloy investigations on the influence of vibrational treatment during solidification were performed in cooperation with a heavy metal foundry and the effects of the vibrational treatment on the properties of the casting were evaluated.

Selected castings revealed significant grain refinements that led to improved properties. Proposals to transfer the procedure to industrial practice were elaborated under the supposition that the vibrational device can be integrated into the existing pouring line and that:

- The tensile stress (strength) increases with frequency.
- In the frequency range considered, vibration decreases the ductility of the material as evidence from the percentage elongation and percentage reduction in area.

- In the frequency range considered, vibration increases the number of grains formed, that is smaller grains and hence fine grain structure. This confirms that increasing copper contents increases strength and hardness while ductility decreases.

REFERENCES

1. Murty, H. and Donnelly, C. 2002. 32nd International Annual Conference of the ICT. 91
2. Easton, P. 2000. “Dynamic Triangulations for Granular Media Simulations”. *J. of Iron and Steel Institute*.
3. McKay, Johnson, et al. 1993. *Selection and use of Engineering Materials. 1st Edition*. Schumacher and Greer Butterworths: London, UK
4. Mohanty and Gruzleski. 1995. *Physical and Engineering Properties of Cast Metals*. The British Cast Iron Research Assn: Birmingham, England.
5. Vandyoussefi, E., Tronche, A. et al. 1961. *ASM Metals Handbook. 8th Edition, Vol 1*. America Society for Metals: Metals Park, OH. 349-365.
6. Banerji, R., Delamore, W., and Morrogh, H. 1968. “The Statues of the Metallurgy of Cast Iron”. *Journal of the Iron and Steel Institute*. 1-10.
7. Rehder, J.E. 1965. “The Critical Temperature Range in Cast Irons”. *Transactions, American Foundry Men’s Society*. 73:473- 487.
8. Fisher, T. F. 1973. *Journal of the Inst. Of Brit. Foundation*. Column 66, Part 3.
9. Baladin, G. .F and Yakov Lev, P. 1964. *Russian Casting Production Methods*: MIR Publishers: Moscow, USSR.
10. Contrells, A. 1975. *Solids, Liquids and Solidification: An Introduction to Metallurgy*. Edward Arnold: London, UK. 165.
11. Bounds, G.M. 1958. *Nucleation in the Solidification of Metals, Liquid Metals and Solidification*. American Society for Metals: Cleveland, OH. 187.
12. Jackson, K.A. 1958. *Mechanism of Growth, Liquid Metals and Solidification*. American Society of Metals: Cleveland, OH. 187.
13. Davies, G. J. 1973. *Nucleation Growth: The Structure of Casting and Defects in Casting*,

Solidification and Casting. Applied Publishers, Ltd.: London, UK. 12-28, 30-39, 75-115, and 166-185.

14. Flemings, M. C. 1975. Solidification of Casting and Ingots, Nucleation and Interface Kinetics: Solidification Processing. McGraw- Hill: New York, NY. 134-172 and 290-319.
15. Reed Hill, R.E. 1968. *Solidification of Metals: Physical Metallurgy Principals*. D. Van Nostrand Company: New York, NY. 367-398.
16. Haque, M.M. and Maleque, M. A. 1998. *Journal of Material Processing Technology*. 77:125.
17. Campbell, J. 1991. *Castings*. Butterworth-Heinemann, Ltd: Oxford, UK.
18. Burns, T. A. 1989. *The Fosco Foundryman Handbook. 9th Edition*. Pergamon Press: Oxford, UK.

SUGGESTED CITATION

Adegbuyi, P.A.O., J.O. Uhomoibhi, K.A. Adedeji, and N.A. Raji. 2010. "The Effect of Pouring and Vibration on Cast Quality". *Pacific Journal of Science and Technology*. 11(1):45-54.

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