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Relation between processing, microstructure and mechanical properties of rheocast Al–Cu alloys

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The role of microstructure in modifying the tensile strength of rheocast Al–10 wt % Cu and Al–4.5 wt % Cu is investigated by modelling these products as particulate composites. The tensile strength of rheocast Al–10 wt % Cu alloy is found to depend on experimentally determinable parameter $\bar{D}\bar{Z}$ which is an undefined function of the size distribution of proeutectic α-particles. These products, however, possess lower strength than the conventional casting at equal porosity level. The tensile strength of the rheocast Al–4.5 wt % Cu alloy shows an inverse square root dependence on the proeutectic α-particle size, $\bar{D}$, and therefore appropriate rheocasting process parameters may be employed to bring about grain refinement in order to produce rheocastings possessing higher strength than the conventional casting at equal porosity level.

1. Introduction

Rheocasting is a new casting technique in which a liquid alloy is vigorously agitated during its partial solidification to yield a slurry comprised of non-dendritic spheroidal particles suspended in the remaining liquid. Such a slurry is suitable for making castings of desired shapes by conventional methods [1, 2]. This method of casting is believed to offer many technological advantages such as improvements in die-life and casting quality [3–8], grain refinement without any external grain-refining agent [9–11] and thixoprocessing, using rheocast billets by mechanical processing techniques, e.g. extrusion, at very low deformation loads [12]. In a number of publications [8, 13–21] the mechanical properties of rheocast and thixocast products have been reported, but no detailed study has so far been undertaken to quantify the mechanical properties in terms of microstructural parameters, and therefore the role of rheocast microstructure in modifying the mechanical properties remains uncertain.

The major constituent phases present in hypoeutectic Al–Cu alloys are the aluminium-rich α-solid solution and the α + $CuAl_2$ eutectic. Rheocasting of these alloys permits dispersion of non-dendritic spheroidal and ductile particles of proeutectic α-phase in a matrix of the brittle eutectic [22–24]. Such a constitution of the rheocastings allows us to model them as brittle matrix ductile dispersoid particulate composites. This concept has been utilized in the present investigation to study the mechanical behaviour of rheocast Al–Cu alloys with a view to enhance our understanding of the interrelationship of processing, microstructure and mechanical properties of these products. Two compositions, Al–4.5 wt % Cu in the single-phase region, and another, Al–10 wt % Cu, in the two-phase region of the Al–Cu phase diagram, have been selected for making the rheocastings. The tensile strength of these castings have been related quantitatively to the microstructural parameters and their dependence on the processing condition has been examined.

2. Theoretical considerations

In forming a theoretical basis for the strength of the type of composite considered above, two situations must be recognized. The first situation corresponds to the case when the dispersed α-phase is present as discrete particles and the eutectic matrix is continuous. This occurs in the Al–10 wt % Cu alloy in which the volume fraction of the dispersed α-particles is low. The second situation corresponds to high volume fraction, when the α-particles join each other to form a continuous network and the eutectic becomes discontinuous as in the case of Al–4.5 wt % Cu alloy. When a composite of the former type is deformed under tension, the force is not applied directly to the particles, but it is transferred through the matrix. For such a case, a shear-lag analysis has been developed by the present authors [24] which relates the composite strength, σ_{uc}, to the matrix strength, σ_{um},

$$\sigma_{uc} = \sigma_{um}\left[V_m + \frac{4V_p}{\pi^2}(\bar{D}\bar{Z})^2\right] \qquad (1)$$

where V_m and V_p are volume fractions of matrix and particles, respectively, $\bar{D}$ is the average particle size obtained from measurements of particle section size on a planar section, and $\bar{Z}$ is the average of reciprocals

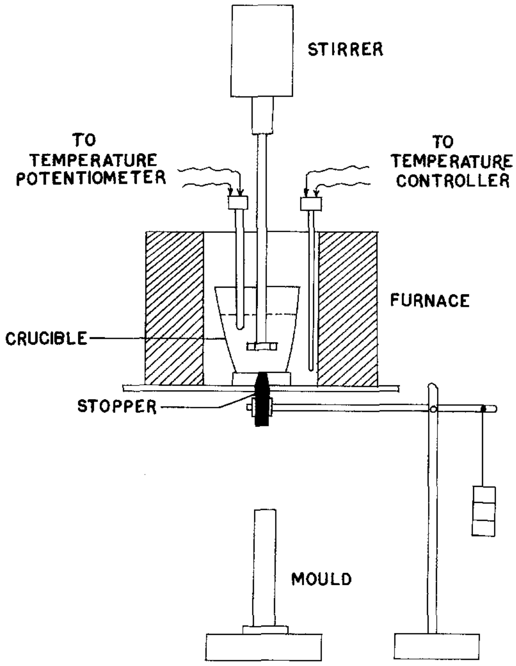

Figure 1 Experimental apparatus for rheocasting.

of these measurements. In the development of Equation 1 it has been assumed that both the matrix and the particles are elastically deformed, the particle–matrix interface is strong enough to effect load transfer, the matrix and the particles have equal shear moduli, Poisson's effect is negligible, and the failure of the matrix leads to the failure of the composite.

For the case when the α-phase is continuous and the eutectic is discontinuous, the force is not applied directly to the eutectic but it is transferred through the α-phase when the composite is deformed in tension. Thus the roles of the α-phase and the eutectic are reversed. The general equation for the composite stress, σ_c, follows directly from Equation 1

$$\sigma_c = \sigma_\alpha (V_\alpha + \beta V_e) \quad (2)$$

where σ_α represents the stress in the α-phase, V_α and V_e are the volume fractions of α-phase and eutectic, respectively, and β is a factor determining the extent of stress shared by the eutectic and depends upon its geometry and distribution. The geometrical features and the distribution of the eutectic are not precisely defined, and therefore a unique estimate of β is not possible. However, as discussed later, Equation 2, is useful in the analysis of the mechanical behaviour of such composites.

3. Experimental procedure

The details of the experimental arrangement for making rheocastings is shown schematically in Fig. 1. Both the alloys, Al–4.5 wt % Cu and Al–10 wt % Cu were prepared from commercial purity aluminium. For each casting, about 500 g alloy was melted in the graphite crucible having a 12 mm hole at the bottom plugged with a graphite stopper. After melting, the furnace was switched off and the melt was allowed to cool inside the furnace while the melt temperature was continuously measured with a sheathed chromel–alumel thermocouple using a temperature potentiometer. When the temperature came down to the liquidus, a stirrer having a four-blade impeller was introduced and the melt agitated vigorously during its primary solidification. At the desired pouring temperature, stirring was stopped, the graphite stopper removed and the slurry was cast into a 30 mm × 30 mm × 250 mm laboratory-size mould. Suitable samples were machined out from each casting for metallographic examinations and tensile tests. Some castings were also made in the conventional manner for comparison with rheocastings. Typical microstructures of rheocastings and conventional castings of the two alloys are shown in Figs 2 and 3. Tensile tests were performed on a Hounsfield tensometer. Before the tensile test, each specimen was evaluated for its porosity content by the weight-loss method. Three specimens for each casting were tested and the average of the three constituted one reading. The size and distribution of α-phase particles were determined by standard techniques of quantitative metallography.

4. Results and discussion

4.1. Al–10 wt % Cu alloy

Typical load–extension curves recorded during the tensile test are depicted in Fig. 4. The curve for the Al–10 wt % Cu alloy shows that the deformation of this alloy is elastic up to fracture and therefore the strength of this alloy is primarily controlled by the brittle eutectic matrix. Using Equation 1 for the microstructural contribution, and the model of Ghosh

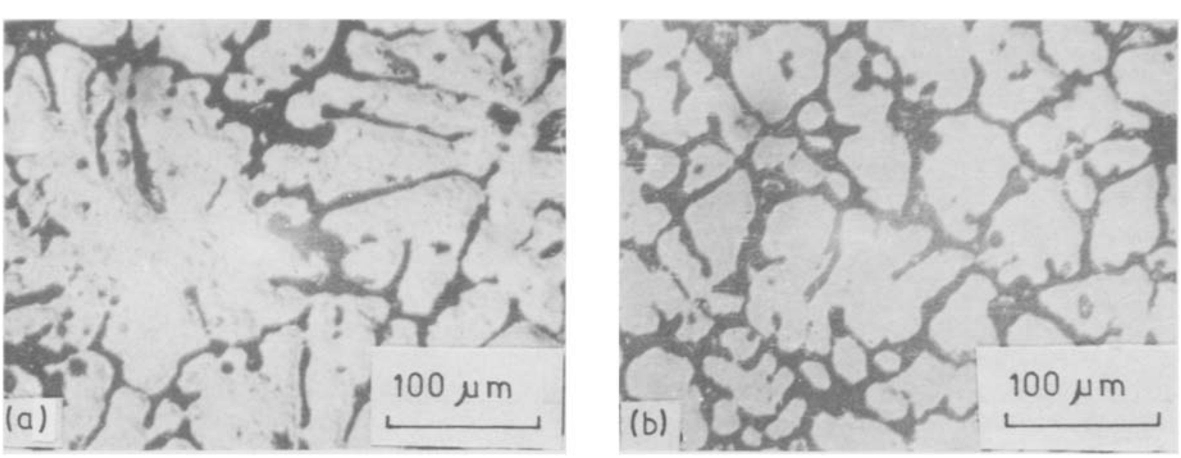

Figure 2 Microstructure of Al–10 wt % Cu alloy (a) conventionally cast, (b) rheocast.

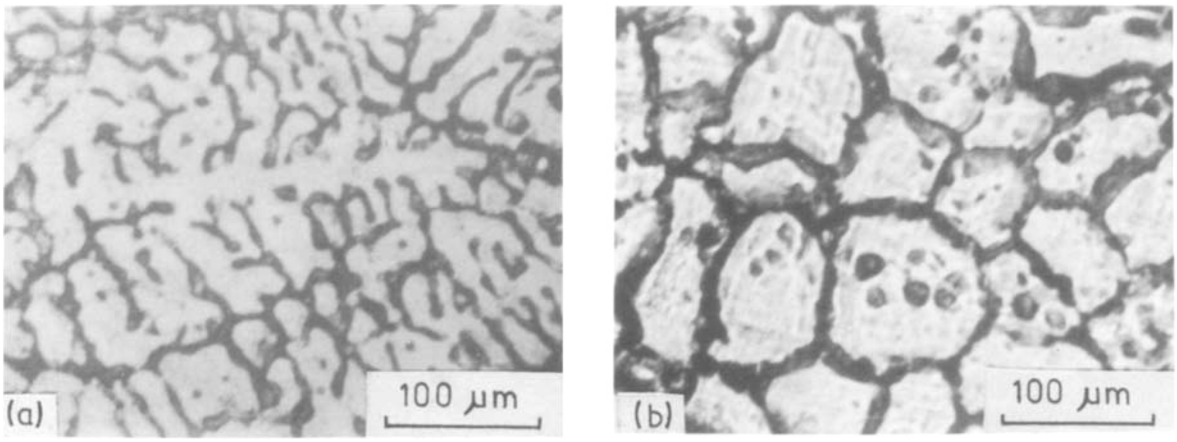

Figure 3 Microstructure of Al–4.5 wt % Cu alloy (a) conventionally cast (b) rheocast.

et al. [28] for linear negative contribution of porosity in stir-cast composites, the following correlation equation is obtained for the tensile strength from the least square fit of the experimental data

$$\sigma_{uc}\,(\mathrm{MN\,m^{-2}}) = 75.0 + 67.97\,(\bar{D}\bar{Z})^2 - 4.71\,p \quad (3)$$

where p is the porosity (%). Equation 3 shows good agreement with the experimental observations within $\pm 10\%$ of deviation [24].

For making an assessment as to how the strength of conventional casting compares with that of rheocasting because of microstructural modifications brought about by the rheocasting process, the strength values must be examined at equal porosity levels. The conventional casting made for this purpose shows a porosity level of 6.3%. Using Equation 3, the strength values of various rheocastings are calculated for a porosity level of 6.3% and plotted against the particle parameter $(\bar{D}\bar{Z})^2$ in Fig. 5. The strength level of conventional casting is also shown in this figure. It is observed that for the $\bar{D}\bar{Z}$ values obtained in the present investigation, all the rheocastings have lower strength than the conventional casting at equivalent porosity levels. Fig. 5 also suggests that at higher $\bar{D}\bar{Z}$ values, the rheocastings approach the strength level of conventional casting. It follows, therefore that in order to maximize the strength of rheocastings, the process variables should be so selected that $\bar{D}\bar{Z}$ is maximum.

The relation between the parameter $\bar{D}\bar{Z}$ and the process variables is not precisely known. However, it may be shown that the parameter $\bar{D}\bar{Z}$ represents the ratio of arithmetic mean to harmonic mean of the particle size measurements. This ratio is unity for uniform-sized particles and greater than unity if particles of varying sizes are present. It follows, therefore, that $\bar{D}\bar{Z}$ is dependent upon the particle-size distribution. Although a unique relation between $\bar{D}\bar{Z}$ and particle-size distribution does not exist, a plot of $\bar{D}\bar{Z}$ against coefficient of variation (ratio of standard deviation to average particle size) in Fig. 6 is revealing. The figure suggests that $\bar{D}\bar{Z}$ increases with increase in the coefficient of variation. An earlier result [23] shows that the coefficient of variation is not significantly affected by the stirring speed, but it increases with increase in the pouring temperature. Thus, a higher pouring temperature is required for obtaining higher $\bar{D}\bar{Z}$ values and consequently higher tensile strength. This observation is significant in that it suggests that the strength is maximum for the processing condition where the effect of primary solidification is minimum and maximum opportunity is available for the dendritic solidification in the mould. Such a processing condition approaches the condition of conventional casting, and therefore a tensile strength higher than the conventional casting is not obtainable in case of Al–10 wt % Cu alloy through the microstructural modifications brought about by the rheocasting process.

The published literature does not provide any information on the subject to enable comparison with the present findings. However, it is pertinent to examine the model of Evans *et al.* [25] for the brittle matrix–ductile particle composites. It has been proposed that the role of ductile particles is to restrain the crack propagation through the brittle matrix. The

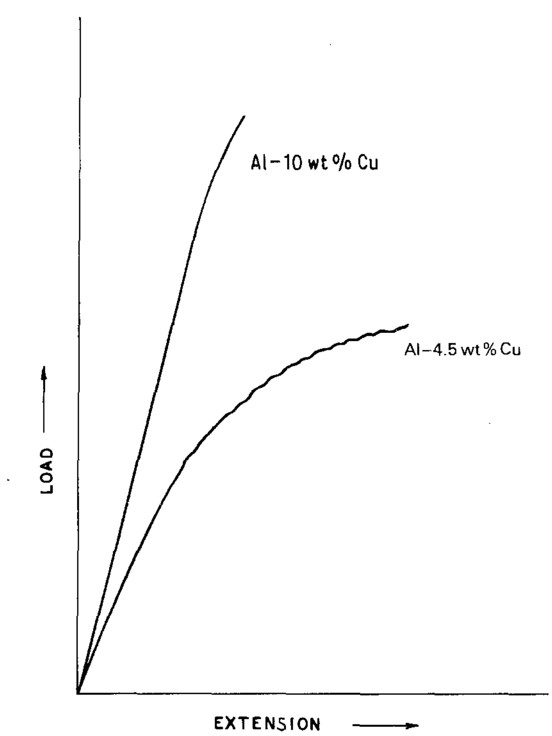

Figure 4 Typical load–extension curves recorded during tensile testing.

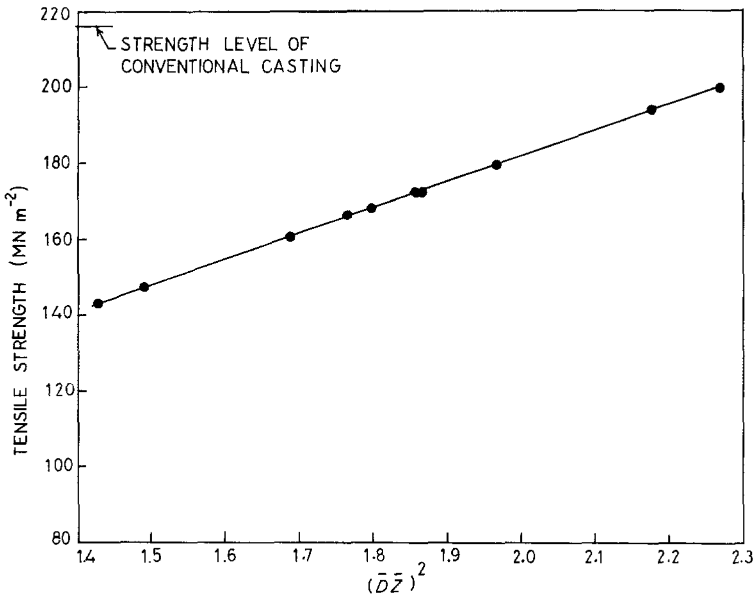

Figure 5 Variation of tensile strength with $(\bar{D}\bar{Z})^2$ in rheocast Al–10 wt % Cu alloy at 6.3% porosity level.

mechanism by which such a restraint is affected is believed to be compressive stresses which the ligaments of unbroken particles joining the crack faces exert to restrain the displacement of crack faces. It has been suggested that for the above mechanism to be operative, particles of cylindrical morphology should be used instead of spheroidal particles. When spheroidal particles are employed, the crack in the matrix simply by-passes the particles and no crack particle interaction results. It appears, therefore, that the low strength of rheocasting containing spheroidal particles is related to the inability of α-particles to restrain crack propagation through the brittle eutectic matrix.

4.2. Al–4.5 wt % Cu alloy

In contrast to the load–extension curve for the Al–10 wt % Cu alloy, the curve for the Al–4.5 wt % Cu alloy depicted in Fig. 4 shows significant plastic deformation associated with serrations in the flow curve which continues until the test piece fractures. This suggests that the ductile α-phase has been deformed to fracture and the fracture of the α-phase has led to the ultimate failure of the alloy.

The general expression for the composite stress, $\sigma_c(\varepsilon)$ as a function of strain, ε, may be written from Equation 2 as

$$\sigma_c(\varepsilon) = K\varepsilon^{1/2}(V_\alpha + \beta V_e) \quad (4)$$

where $K\varepsilon^{1/2}$ represents the flow stress of α-phase (σ_α) as a function of strain. The stress corresponding to the start of serrated flow, $\sigma_c(\varepsilon_s)$, may then be written as

$$\sigma_c(\varepsilon_s) = K\varepsilon_s^{1/2}(V_\alpha + \beta V_e) \quad (5)$$

where ε_s is the strain at which the serrated flow begins. Similarly the composite stress at fracture, $\sigma_c(\varepsilon_f) = \sigma_{uc}$,

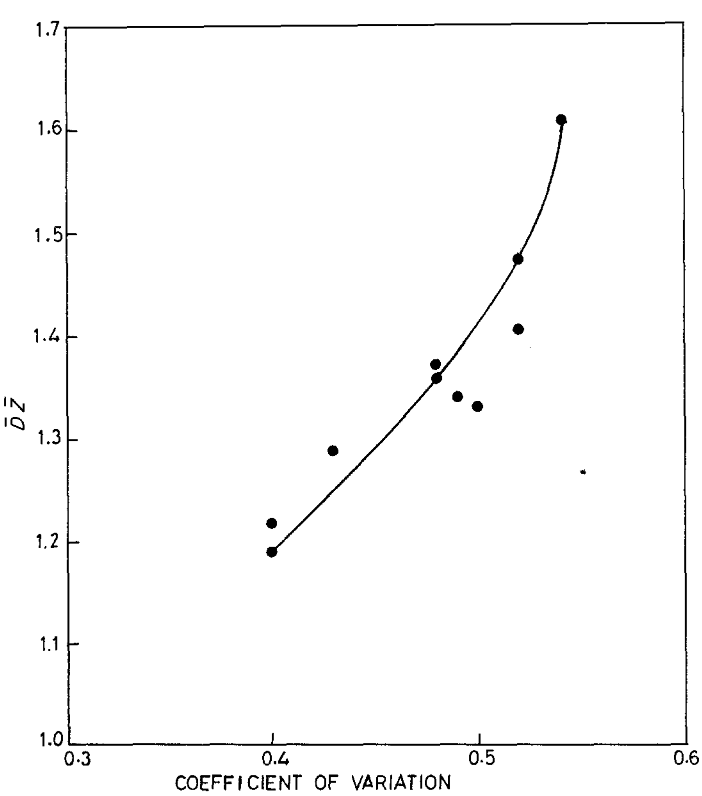

Figure 6 Coefficient of variation plotted against $\bar{D}\bar{Z}$.

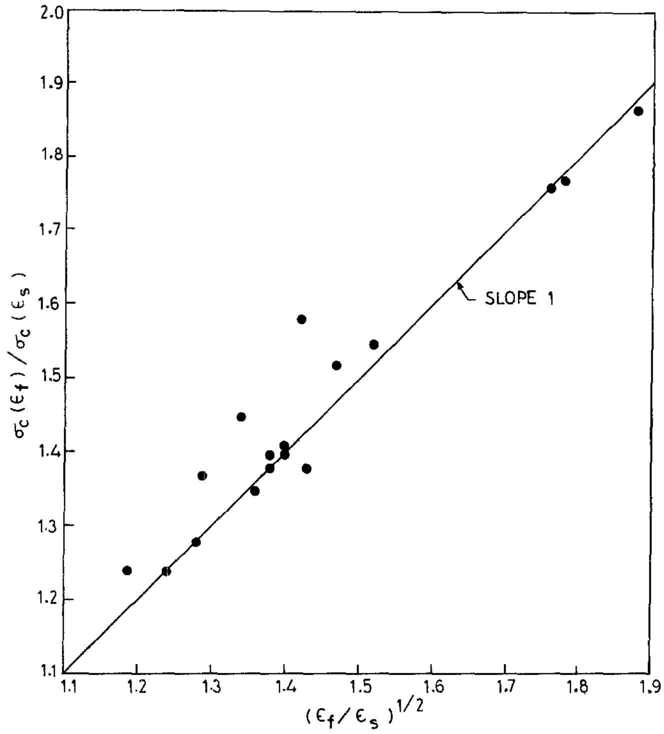

Figure 7 Relationship between $\sigma_c(\varepsilon_f)/\sigma_c(\varepsilon_s)$ and $(\varepsilon_f/\varepsilon_s)^{1/2}$ according to Equation 7 for rheocast Al–4.5 wt % Cu alloy.

as a function of fracture strain, ε_f, is given by

$$\sigma_c(\varepsilon_f) = K\varepsilon_f^{1/2}(V_\alpha + \beta V_e) \quad (6)$$

The combination of Equations 5 and 6 yields

$$\frac{\sigma_c(\varepsilon_f)}{\sigma_c(\varepsilon_s)} = \left(\frac{\varepsilon_f}{\varepsilon_s}\right)^{1/2} \quad (7)$$

On the basis of experimentally observed values of ε_f, ε_s, $\sigma_c(\varepsilon_f)$ and $\sigma_c(\varepsilon_s)$, a plot of $\sigma_c(\varepsilon_f)/\sigma_c(\varepsilon_s)$ against $(\varepsilon_f/\varepsilon_s)^{1/2}$ has been made in Fig. 7. It is observed that the experimental points are well distributed around the line with the slope of 1. Therefore, the strength of this alloy is primarily α-phase-controlled and the validity of Equation 2 is confirmed.

In most metals and alloys, a Hall–Petch-type relation has been extensively used to describe empirically the dependence of fracture strength or tensile strength on particle size. Using a similar relation, the least square fit of the experimental data yields the following correlation equation

$$\sigma_{uc}\,(\mathrm{MN\,m^{-2}}) = 159.3 + 81\,(\bar{D})^{-1/2} - 6.15\,p \quad (8)$$

For various rheocastings, the strength values calculated according to Equation 8 have been compared with the experimental values in Fig. 8. It is observed that the agreement between the calculated and the experimental strength values is generally good within $\pm 10\%$ deviation. The conventional casting made for comparison with rheocastings, shows a porosity level of 1.5%. The strength values of rheocastings are

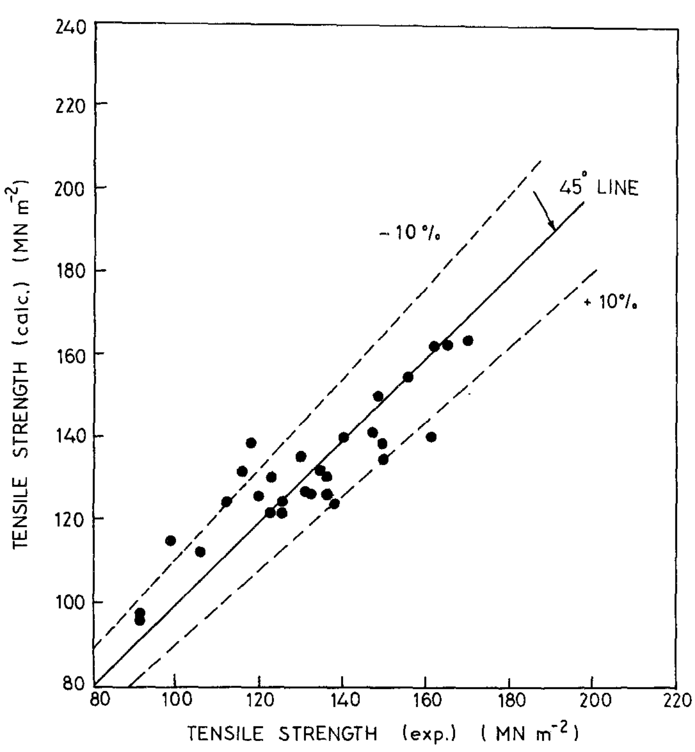

Figure 8 Comparison between experimental and calculated (according to Equation 8) tensile strength values of rheocast Al–4.5 wt % Cu alloy.

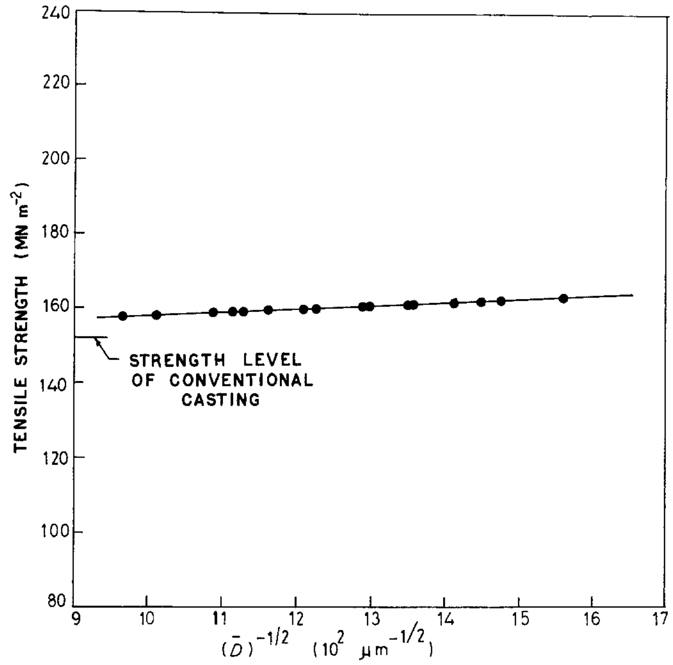

Figure 9 Variation of tensile strength with α-particle size in rheocast Al–4.5 wt % Cu alloy at 1.5% porosity level.

calculated at 1.5% porosity level using Equation 8 and are plotted in Fig. 9. The strength level of conventional casting is also indicated in this figure. It is apparent that the rheocastings have higher strength than the conventional casting at equal porosity level, although the difference is not very significant. However, this curve suggests that an improvement in strength is possible if appropriate process parameters are selected to produce extremely fine size of α-particles. In order to achieve this objective, a high stirring speed and high pouring temperature may be considered as appropriate processing conditions [26, 27]. However, a high pouring temperature offers greater opportunity for the dendritic solidification in the mould and a lower pouring temperature results in larger particle sizes due to growth and particle coalescence. The production of an extremely fine rheocast microstructure calls for optimization of the process parameters.

5. Conclusions

On the basis of the present investigation, the following conclusions can be drawn.

1. The tensile strength of rheocast Al–10 wt % Cu alloy is governed by an experimentally determinable parameter $\bar{D}\bar{Z}$ which increases with the size distribution of proeutectic α-particles.

2. The rheocast Al–10 wt % Cu alloys have a lower tensile strength than conventional castings at equal porosity level, and therefore rheocastings possessing higher tensile strengths than conventional castings are unlikely to be produced through microstructural modifications brought about by the rheocasting process.

3. The tensile strength of rheocast Al–4.5 wt % Cu alloy has an inverse square root dependence on the proeutectic α-particle size, $\bar{D}$.

4. The rheocast Al–4.5 wt % Cu alloys have equivalent or slightly higher tensile strengths than conventional castings at equal porosity level. Suitable processing conditions may be selected to refine the α-particle size in order to produce rheocastings possessing higher tensile strengths than conventional castings.

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