Letters

A relationship between dendrite arm spacing and atomized powder diameter

An expression between secondary dendrite arm spacing, d, and atomized powder diameter, D, can be obtained from the equation

$$d = b(\dot{T})^{-n} \tag{1}$$

in which b and n are empirical constants and $\dot{T}$ is the cooling rate. Equation 1 has been applied to quenched alloys by taking $\dot{T}$ as either the local average cooling rate during solidification or just the cooling rate [1] and we consider that it also applies to powder atomization processes. For this treatment $\dot{T}$ is taken as the cooling rate at the liquidus temperature, T_L, and is obtained from the standard heat transfer equation for the convective and radiative cooling of a sphere which contains negligible temperature gradients:

$$\dot{T} = \frac{6}{D\rho C_p}\,[h_c(T_L - T_0) + \epsilon\sigma(T_L^4 - T_0^4)]\,. \tag{2}$$

In Equation 2 ρC_p is the heat capacity per unit volume of the atomized droplet and T_0 is the temperature of the surrounding medium, while ϵ, σ and h_c are the emissivity of the powder, the Stephan–Boltzmann constant and the convection heat transfer coefficient, respectively.

All of these quantities should be constant for a given alloy system except for h_c which depends on both the atomization process and the powder diameter. This variation for a sphere moving through a gas under turbulent flow conditions, is given by [2]:

$$h_c = \frac{ak_f}{D}\left(\frac{Dv\rho_g}{h_f}\right)^{0.6} \tag{3}$$

in which k_f and h_f are the conductivity and dynamic viscosity of the gas at the convective fibre temperature, ρ_g and v are the density and the velocity of the powder relative to the gas and a is a constant which depends on the system of units. The powder velocity, v, should also be a function of the powder diameter, since by increasing the energy expended in an atomization process one expects both that the powder size will decrease and its velocity will increase. This was proven also from our experimental work. For example, if in the rotating electrode process [1] one equates the surface tension force tending to hold a liquid droplet to the electrode to the centrifugal force tending to throw it off, one obtains that v is inversely proportional to D.

We thus can accept for all atomization processes that

$$h_c = \lambda D^{-m} \tag{4}$$

in which λ should be approximately constant for a given process and material being solidifed and m could be obtained either theoretically or from the experimental data. Combining Equations 1, 2 and 4 yields the final result:

$$d = b\left(\frac{D\rho C_p}{6}\right)^n\left[\frac{\lambda}{D^m}(T - T_0) + \epsilon\sigma(T^4 - T_0^4)\right]^{-n} \tag{5}$$

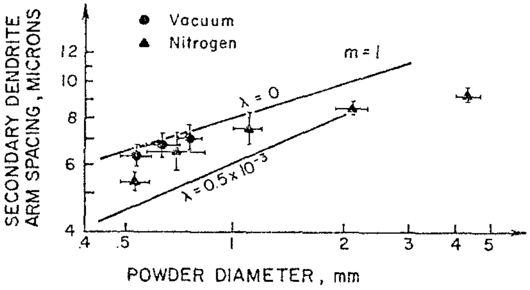

Figure 1 Secondary dendrite arm spacing versus diameter of spherical powders of maraging 300 alloy in vacuum and nitrogen.

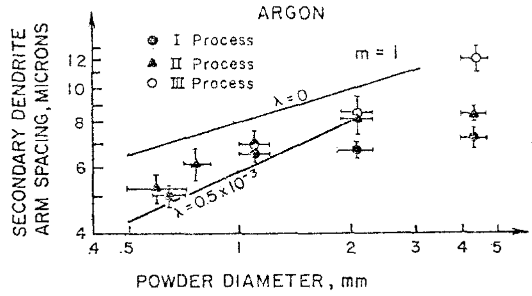

Figure 2 Secondary dendrite arm spacing versus diameter of spherical powder of maraging 300 alloy in argon.

A comparison between Equation 5 and the data of Joly and Mehrabian for $m = 1$ and for maraging 300 alloy [4] is given in Figs. 1 and 2. For the comparison, the values [1] of $b = 39.8$, $n = 0.3$ and $T = 1773$ K were used. The liquidus temperature values for Fe–25% Ni and reported values of $\epsilon = 0.33$ and $\rho C_p = 1.27 \text{ cal cm}^{-3} \text{ K}^{-1}$ of the same alloy were also used [3]. The atomization processes I, II and III [1] are carried out at 1732, 1563 and 1554°C tap temperatures while the side gas nozzle gauge pressures are 3.4×10^5, 4.1×10^5 and $4.8 \times 10^5 \text{ N m}^{-2}$ respectively (or 50, 60, 70 psi) in argon.

When the powder size increases, the data show a deviation from the dendrite arm spacing expected for large size powders, although there is a need for more data in that area. In Figs. 1 and 2 the $m = 1$ rough guess for the rotating electrode process is plotted for powders of diameter less than 2 mm for a comparison against experimental points of different atomization processes.

Of interest in Equation 5 is the prediction that d is proportional to D^n for radiation cooling and proportional to $D^{-n(m-1)}$ when convection cooling dominates. If a process were modified to increase λ, for example by increasing the environmental gas pressure, one would predict that even smaller dendrite arm spacing could be achieved. For example, by doubling the gas pressure, λ would be increased by a factor 1.5 and the effect it would have on dendrite arm spacing can be seen in Figs. 1 and 2.

Fig. 1 contains plots of equation 5 in logarithmic scale for $m = 1$ for $\lambda = 0$ and $\lambda = 0.5 \times 10^{-3} \text{ cal cm}^{-1} \text{ sec}^{-1} \text{ K}^{-1}$ and experimental points for vacuum and nitrogen while Fig. 2 contains experimental points for argon and for three different atomization processes.

Acknowledgement

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References

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Deformation of single crystals of bis (p-toluene sulphonate) diacetylene polymer

Thermal polymerization of single crystals of bis (*p*-toluene sulphonate) diacetylene produces large highly perfect polymer single crystals [1–4]. Such crystals will have highly anisotropic elastic properties; parallel to the chain the modulus will be large but perpendicular to the chain it will be much smaller. Deformation perpendicular to the polymer chain will depend on the packing of the sidegroups as well as the flexural properties of the backbone. The bis (*p*-toluene sulphonate) polymer has a well defined system of stress twining when subjected to forces perpendicular to the polymer chain [2, 4]. However, the deformed regions cannot be true twins since the polymer chains must be packed differently [4].

The bis (*p*-toluene sulphonate) polymer is monoclinic with symmetry $P2_1/b$ in the first setting (or $P2_1/c$ in the second setting) with two polymer chains traversing the unit cell [5]. The deformation twins have slip traces along $[1\,0\,0]$, $[1\,0\,1]$ and $[\bar{1}\,0\,1]$ on the $(0\,1\,0)$ crystal facet [4] and along $[1\,2\,0]$, $[1\,\bar{2}\,2]$ and $[1\,2\,2]$ on the $(2\,1\,0)$ facet [2]. Transmission microscopy shows that the $[1\,0\,0]$ $(0\,1\,0)$ and $[1\,2\,0]$ $(2\,1\,0)$ traces result from deformation on the $(0\,0\,1)$ plane while the other traces correspond to the inclined planes $(3\,2\,2)$ and $(3\,2\,\bar{2})$. The deformed regions are displaced almost normal to the $(0\,1\,0)$ facet, i.e., along $[1\,2\,0]$ in the $(0\,0\,1)$ plane and along $[1\,2\,4]$ and $[1\,2\,\bar{4}]$ in the inclined planes, all these directions join nearest neighbour polymer chains in successive $(0\,1\,0)$ planes.

Projections normal to the $(3\,2\,2)$ and $(3\,2\,\bar{2})$