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 $\rho C_p = 1.27$ cal cm^{-3} 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 4.8×10^5 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 $\lambda = 0.5 \times 10^{-3}$ cal cm^{-1} sec^{-1} K^{-1} and experimental points for vacuum and nitrogen while Fig. 2 contains experimental points for argon and for three different atomization processes.

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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 $\bar{2}$)

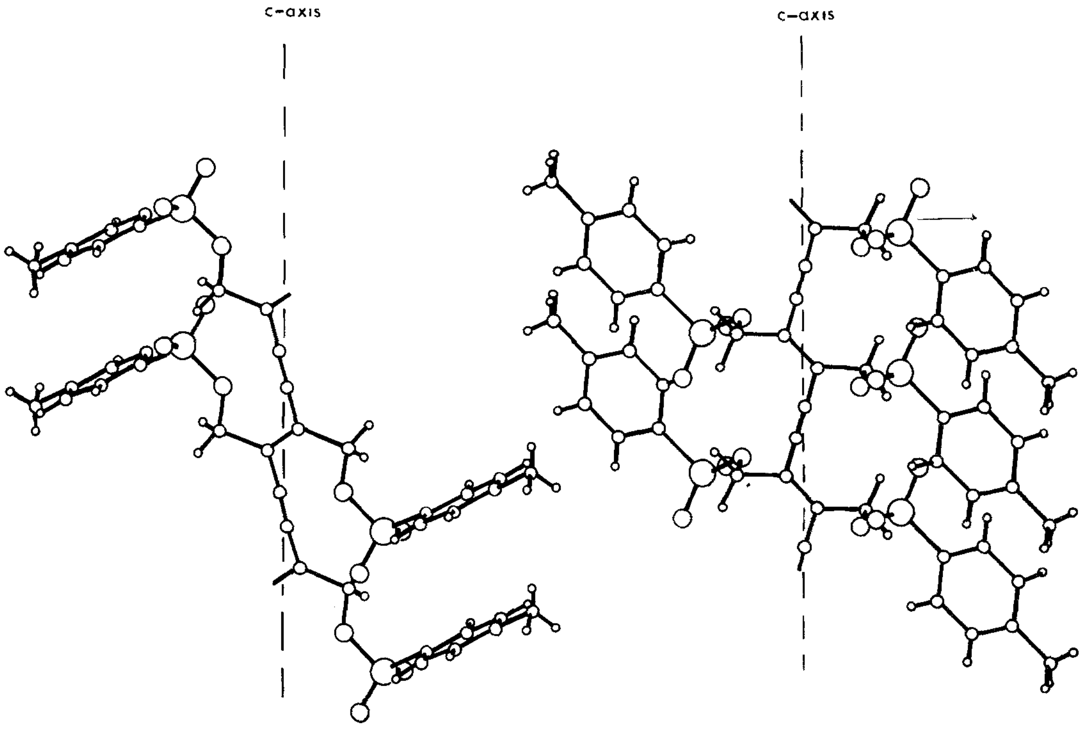

Figure 1 Polymer chains of bis (*p*-toluene sulphonate) diacetylene polymer viewed along the [1 2 4] direction. The polymer chains are shown separately to avoid confusion due to overlap of sidegroups.

planes were obtained using the structure of Kobelt and Paulus [5]. These show that the planes of the toluene rings in the side groups lie in the (3 2 2) plane for one of the chains in the unit cell and in the $(3\,2\,\bar{2})$ for the other, see Fig. 1. Thus, along these planes bending of one of the polymer chains is made relatively easy by sliding of the side groups in the plane of the toluene rings. Deformation on these planes is, however, rendered difficult by the second less favourably orientated chain. Deformation on the (0 0 1) plane, which is symmetrically disposed between (3 2 2) and $(3\,2\,\bar{2})$, requires a similar deformation of both chains. Thus, the occurrence of three distinct deformation systems is a result of the packing of the side groups.

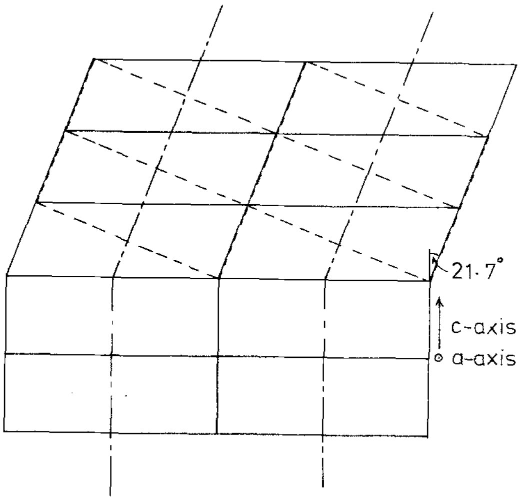

Figure 2 Section of a bis (*p*-toluene sulphonate) diacetylene polymer crystal deformed on the (0 0 1) plane viewed along [1 0 0]. The original and distorted unit cells are drawn with full lines and the new monoclinic cell in the deformed region with dashed lines. Polymer chains are shown by the chain lines.

In most cases the deformed regions are very small [4] and a precise measurement of the orientation of the deformed regions is not possible. However, one sample cooled to liquid helium temperature for spectroscopic observations developed deformed regions along (1 0 0) on a (0 1 0) facet about 0.2 mm wide and 1 cm long. The orientation of the deformed facet was measured using an optical goniometer and was found to be 21.7 ± 0.1° away from the plane of the undeformed (0 1 0) facet. The lattice distortion is shown in Fig. 2, the original unit cell becomes triclinic but a new monoclinic cell can be defined as shown in the figure. The *a*-axis is unchanged but the *b*-axis is reduced from 14.93×10^{-10} to 13.7×10^{-10} m and γ is decreased from 118.14° to 119.9°. The dimensions of the polymer backbone can be determined from the resonance Raman spectrum, since only vibrations of the backbone are observed and these are sensitive to changes of bond lengths and angles [3, 6–8]. The spectra of the deformed and undeformed regions were identical, thus the *c*-axis must be unchanged in the new cell. The new cell has a 7% smaller volume than the original cell so that there must be considerable rearrangement of the side groups. Further studies are in hand in order to determine the form of the distortion for the other deformation system.

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Hardness and constitution in rhodium–cobalt alloys

This note describes some experiments which were carried out to provide background for an investigation [1] into some physical properties of rhodium-rich rhodium–cobalt alloys. In view of the fragmentary nature of the information, it would be inappropriate to be other than brief. However, the reader may like to be reminded that the rhodium–cobalt system is apparently simple, with complete solid solubility between rhodium and the high-temperature phase of cobalt. Interest arises from the allotropic change in cobalt, and its temperature variation with alloying; the transformation temperature, after rising to a maximum near 16% rhodium, decreases steadily to approximately room temperature at 50% rhodium (the highest rhodium content studied by Köster and Horn [2, 3]). It was not unreasonable to expect possible martensitic change at subzero temperatures in rhodium-rich alloys, and hardness measurement was, therefore, a convenient tool for an investigation. Essentially, then, samples, which actually span a somewhat wider range of composition than has been indicated, have been subjected to low temperatures and the hardness variations used as indicators of structural change.

Alloy buttons weighing less than 1 g were prepared in an argon arc furnace from metals of approximately 4N purity. The buttons were tested using a Vickers microhardness attachment with a load of 200 g; they were then sealed in evacuated silica capsules, homogenized for 16 h at 1150° C, water-quenched and re-tested. Subsequently, the buttons were immersed for 5 min in liquid nitrogen, then in liquid helium, and hardness tested after each cooling treatment. The results are presented in Fig. 1, where each inset shows a graph, for a single composition, of the hardness (range 200 to 400 kg mm^{-2}) for each of the four heat-treatment states.