at an average distance of 0.6 mm from the interface with the parent metal. A similar but smaller peak was commonly noted in the magnesium content, whereas the zinc content usually showed a minimum value, all values being recorded at the same distance from the interface. Solution treatment of the specimens at 480° C after welding effectively removed these transients in the composition profiles of silver, magnesium, and zinc (Fig. 2b).

The variations in the solute contents that have been described are not periodic in character and it is suggested that the peak values in silver and magnesium arise from a sudden increase in the solidification rate that is to be expected when the welding electrode retreats from a molten zone. The greater magnitude of the peak in silver content when compared with that observed for magnesium is presumed to arise from preferential segregation of the higher melting point element at the solid/liquid interface. The minimum value in zinc content in the same region is a curious result and may perhaps be due to localized volatilization of zinc during arc-welding.

Reference has been made to the possibility of solute banding having an influence on the mechanical properties of welds [5]. The presence of such bands would also be expected to alter the local electrochemistry of a region. Accordingly it has been proposed that this may be one reason why the addition of silver to filler wires has been found to increase the resistance of adjacent regions in the parent metal to intercrystalline cracking [7].

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Determination of the effective size of germanium particles in an Al–Ge alloy with small- and wide-angle X-ray methods

Precipitation phenomena in Al–Ge alloys have been recently studied by X-ray and hardness methods [3, 4]. These studies show that the mean size and shape of Ge precipitates depends on the solution treatment and ageing temperatures. Thus, the quenched-in vacancies play an important role in the formation of Ge nuclei and the subsequent growth of the Ge precipitates. The considerably larger atomic volume of the precipitate phase (22.5 $Å^3$/atom), as compared with that of the matrix (16.6 $Å^3$/atom), is responsible for the stressed condition both of the Ge precipitates and of the matrix [4, 5]. It is known that the shape of a X-ray diffraction line profile depends on the particle size and the microstrain components, too.

In the present work Ge precipitates in an Al–4.0 wt% Ge alloy have been studied by small- and wide-angle methods. The specimens were prepared from super-purity aluminium and germanium (99.999%, Koch Light Laboratories, UK). The specimens were homogenized at 450 and 480° C, and then quenched in ice-water. The ageing temperature was 160° C, and ageing times were 390, 1200 and 3000 min. The percentage of germanium

TABLE I Effective sizes of Ge precipitates in Al–4.0 wt% Ge alloy. The solution treatment temperatures are 450 and 480° C, the ageing temperature 160° C.

Ageing time (min)	Mean cross-section radius of Ge precipitates from SAXS measurements (Å)		Effective size of Ge precipitates from 111 reflection (Å)				R.m.s. microstrains at $3\delta = 55$ Å. ($\times 10^{-3}$)	
			$\langle\epsilon_n^2\rangle = 0$		$\langle\epsilon_n^2\rangle = C/n$			
	R_{450}	R_{480}	$\langle D_{\mathrm{eff}}\rangle_{450}$	$\langle D_{\mathrm{eff}}\rangle_{480}$	$\langle D_{\mathrm{eff}}\rangle_{450}$	$\langle D_{\mathrm{eff}}\rangle_{480}$	$\langle\epsilon_3^2\rangle_{450}^{1/2}$	$\langle\epsilon_3^2\rangle_{480}^{1/2}$
390	67	72	61	91	101	169	9·3	7·4
1200	81	86	66	110	121	175	8·5	5·6
3000	91	96	69	121	124	202	7·7	5·5

in a specimen was tested with a fluorecence method, too. All measurements were performed at room temperature. The same specimens were used in the small- and wide-angle measurements.

According to our previous studies the small-angle X-ray measurements can be interpreted in terms of a rod model [3, 4]. The mean radii of the cross-sections of Ge rods can be calculated from small-angle X-ray measurements [3]. The results are given in Table I, where the corresponding small- and wide-angle values are compared with each other.

The diffraction profiles of the 111 and 311 reflections from germanium precipitates were measured using Cu$K\alpha$ radiation monochromated by a quartz crystal. After background scattering subtraction these profiles were corrected for Lorentz-polarization and structure factors. The correction for instrumental broadening was computed by the Stokes method, using an annealed germanium powder standard for all coefficients. The cosine Fourier coefficient, A_n, is the product of a size coefficient, A_n^s, and a distortion coefficient, A_n^{D}:

$$A_n = A_n^s A_n^{\mathrm{D}}. \quad (1)$$

Because multiple orders of the 111 and 311 reflections from germanium were not available, the method outlined by Gangulee was applied on a single profile [1]. In this method the particle size and microstrain components in the Fourier coefficients can be separated, if the functional form of the average microstrain is known.

The low harmonic cosine Fourier coefficients from an hkl reflection can be written as [1]

$$A_n = (1 - nx)(1 = n^2 y_n), \quad (2)$$

where y_n is a function of an average squared microstrain, $\langle\epsilon_n^2\rangle$, and a variable x depends on the average value of the effective particle size, $\langle D_{\mathrm{eff}}\rangle$.

In this study the presence of dislocations has been assumed to be the principal cause of microstrains. Then we can write $\langle\epsilon_n^2\rangle = C/n$ and $y_n = KC/n$ [1]. The average value of x was computed from all valid solutions of Equation 2 for low harmonics. The effective particle size was obtained from $\langle D_{\mathrm{eff}}\rangle = \delta/x$ [1]. The microstrains were determined from Equation 2 by using the average value of x. The values of $\langle D_{\mathrm{eff}}\rangle$ and $\langle\epsilon_3^2\rangle^{1/2}$ for Ge precipitates as a function of ageing time are shown in Table I.

The corresponding values of $\langle D_{\mathrm{eff}}\rangle$ were calculated from the 311 reflections, too. An example of these measurements is shown in Table II, where the results from the 111 and 311 reflections are compared with each other.

The value of $\langle D_{\mathrm{eff}}\rangle$ depends on the orientation relationship of the {111} planes in Ge precipitates. The angles between the {111} planes and Ge rods can be easily calculated [2]. Because one (111) plane is parallel and the other ones are at a given angle to rod axis, $\langle D_{\mathrm{eff}}\rangle_{111}$ is greater than the corresponding effective value of cross-sections of the rods. Especially in the case of the {311} planes $\langle D_{\mathrm{eff}}\rangle_{311}$ is greater than $\langle D_{\mathrm{eff}}\rangle_{111}$ owing to the orientation of the {311} planes in rods.

The present work shows that the precipitation of Ge in Al–Ge alloys can be studied with the small- as well as wide-angle X-ray methods. In the case of rod shaped particles the small-angle X-ray method gives the mean size of the cross-sections, while the value of $\langle D_{\mathrm{eff}}\rangle_{hkl}$ depends on the orientation of the $\{hkl\}$ planes in precipitates. In Fourier analysis of X-ray diffraction profiles the particle size and microstrain components can be separated, too. The accuracy of the Fourier method

TABLE II The values of $\langle D_{eff} \rangle$ for 111 and 311 reflections, when the solution treatment temperature is 450° C.

Ageing time (min)	Effective size of Ge precipitates (Å)				R.m.s. microstrains for 111 reflection at $3\delta = 55$ Å, and 311 reflection at $3\delta = 68$ Å ($\times 10^{-3}$)	
	$\langle \epsilon_n^2 \rangle = 0$		$\langle \epsilon_n^2 \rangle = C/n$			
	$\langle D_{eff} \rangle_{111}$	$\langle D_{eff} \rangle_{311}$	$\langle D_{eff} \rangle_{111}$	$\langle D_{eff} \rangle_{311}$	$\langle \epsilon_3^2 \rangle_{111}^{1/2}$	$\langle \epsilon_3^2 \rangle_{311}^{1/2}$
390	61	78	101	135	9·3	2·8
1200	66	86	121	147	8·5	2·9
3000	69	87	124	150	7·7	2·7

depends on the error arising mainly from inaccurate estimation of the background intensity. Because the total amount of Ge precipitates is relatively low in a specimen, the counting rates in the measurements are not very high. The counting rate of the 311 reflection is lower than that of the 111 reflection, thus the values of $\langle D_{eff} \rangle_{111}$ are more accurate than those of $\langle D_{eff} \rangle_{311}$. Fourier analysis of X-ray diffraction line profiles is a useful method for the study of precipitation in alloys provided that the microstrain and the particle size components in the Fourier coefficients can be separated.

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Anomalous expansion of cadmium carbonate

Cadmium carbonate crystallizes in the space group $R\bar{3}C$ and is isotypic with the calcite group, a family of AXO_3 compounds in which the XO_3 ions are planar. Though many investigators, Ramdohr and Strunz [1], Swanson *et al.* [2] and Graf [3] report the lattice parameters of cadmium carbonate at room temperature, the accuracy of the values is not high. When this work was under progress, Bayer [4] using copper radiation studied the variation in the lattice parameters of cadmium carbonate and evaluated the average coefficients of thermal expansion in the temperature range 20 to 320° C. He made use of reflections with Bragg angles in the range 20 to 40° and reported negative coefficient of expansion along the *a*-direction. No other reports could be found in the literature on the thermal expansion of cadmium carbonate.

The sample used in the present study was supplied by E. Merck, Germany. The powder sample for the study was prepared by filling it in a thin-walled quartz capillary. Using a Unicam 19 cm high-temperature camera, powder photographs were taken with Fe*K* radiation from a Raymax-60 X-ray unit. It was found that due to the fine particle size of the sample, high angle reflections were not sharp. Repeated annealing of the sample at 250° C improved the pattern slightly. With Cu radiation, a considerable overlap of the reflections is observed in the high Bragg angle region. Hence, Fe*K* radiation is preferred. Details of the experimental technique and the method of evaluating the precise lattice parameters and the coefficients of thermal expansion have been described in an