

Under these conditions the substrate becomes the standard, and small errors caused by misalignment, absorption and eccentricity are minimised, providing the substrate and layer are of identical orientation [3].

The separation between the diffracted lines from the substrate and those from the film is then a measure of the composition of the alloy film; the relative line breadths of the diffracted peaks give a measure of either composition spread or of strain in the epitaxial film. Since the difference in angle (2θ) for 6th order 110 reflections between InP and GaP is 4.1° , and a measurement of 0.02° in 2θ is achievable using a chart recording, the precision with which the layer composition can be determined is little worse than that of the calibration standards.

We have shown that the InP-GaP alloy system obeys Vegard's Law and that the diffractometer provides a non-destructive method of determining the composition of epitaxial layers of the alloy on single crystal substrates of known orientation.

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Kirkendall Effect and Diffusion in the Aluminium-Silver System

Diffusion in the aluminium-silver system has already been studied by Heumann and Dittrich [1], and by Heumann and Böhmer [2]. They have looked at both the interdiffusion coefficient and the partial diffusion coefficients in the aluminium-rich region.

In this work the diffusion coefficients have been determined, in the polycrystalline system, for the concentration in the marker interface, which was found to stay in the phase Ag_3Al . The interdiffusion coefficients were also calculated in the aluminium-rich phase and the calculated values were compared with the results of Heumann and Dittrich [1].

The interdiffusion coefficient has been calculated from the concentration penetration curves according to the graphical method of Matano by means of equation [3] of his paper

$$\tilde{D} = -\frac{1}{2t_a} \frac{dx}{dN_{Ag}} \int_{N_{Ag1}}^{N_{Ag2}} x dN_{Ag}, \quad (1)$$

where N_{Ag} is the atomic fraction and t_a is the diffusion time.

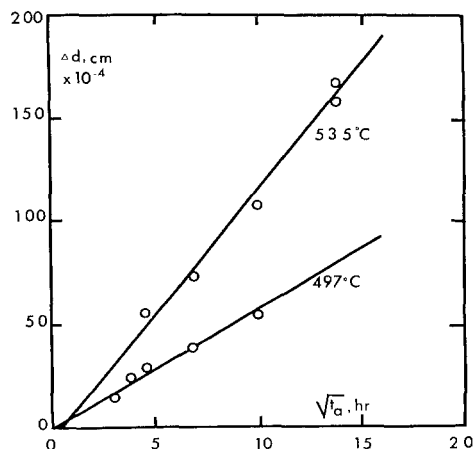


Figure 1 The marker displacement Δd as a function of the square root of the diffusion time t_a .

The partial diffusion coefficients D_{Al} and D_{Ag} can be calculated from the equations according to Darken [3]

$$v = (D_{Ag} - D_{Al}) \frac{dN_{Ag}}{dx} \quad (2)$$

$$\tilde{D} = N_{Ag}D_{Al} + N_{Al}D_{Ag}, \quad (3)$$

where $v = \frac{\Delta d}{2t_a}$ is the velocity of markers.

The metals, aluminium and silver, with a purity of 99.99% were supplied by Koch Light Laboratories and Outokumpu Oy. From this material "sandwich" test-pieces were prepared by welding a platelet of aluminium (dimensions $1 \times 3 \times 10$ mm) to both sides of a platelet

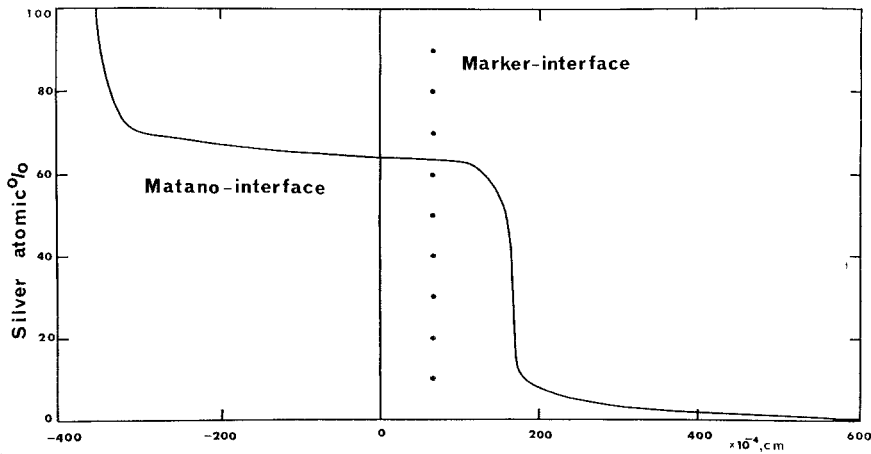
of silver of the same dimensions. Tungsten wires ($\phi = 12 \cdot 10^{-4}$ cm) were used as a marker material. The specimens were annealed in an evacuated tube at the temperatures of 497° C and 535° C. At each temperature 3 to 7 test pieces were annealed simultaneously. Each specimen was used only once.

The distance between the markers was measured by means of a microscope. The marker displacement was calculated from

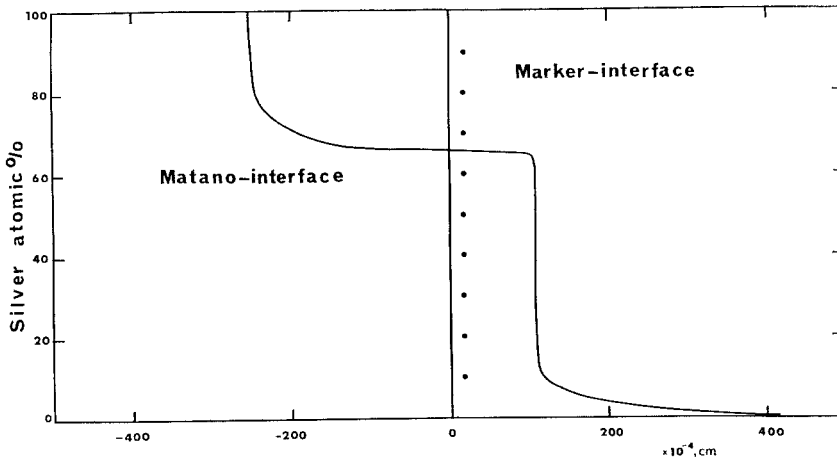
$$\Delta d = \frac{d_a - d_0}{2},$$

where d_0 is the initial distance between the two marker interfaces and d_a is the distance after a diffusion time t_a .

The concentration penetration curves were



(a)



(b)

Figure 2 Concentration penetration curves: (a) 96.0 h at 497° C; (b) 21.5 h at 535° C.

determined by means of microprobe analysis. The specimens for the calibration of micro-analysis curves were prepared by quenching drops of melted alloy in liquid nitrogen.

In each couple the markers were found to move toward the aluminium. The displacements of the marker interface are shown in fig. 1 as a function of the square root of the diffusion time t_a . The value of each point in this figure was obtained by measuring the displacements from 3 to 7 different test pieces.

The concentration penetration curves are shown in fig. 2. The diffusion coefficients \tilde{D} , D_{Ag} and D_{Al} were calculated with the aid of these curves and the results are presented in table I.

Porosity was observed in very few specimens. An electrolytically polished surface, in which the porosity appears in the aluminium-rich region, is shown in fig. 3.

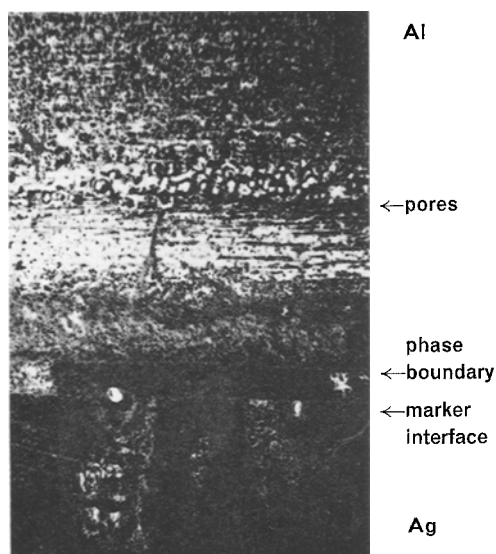


Figure 3 Porosity developed in the aluminium side of an aluminium-silver couple after diffusion for 21.5 h at 535° C in vacuum ($\times 75$).

The partial diffusion coefficients and the direction of the displacement of markers show that the partial diffusion coefficient of aluminium in aluminium-silver alloy containing about

65 at. % silver is higher than the partial diffusion coefficient of silver. It is interesting to compare our results with those of Heumann and Dittrich. They have shown that silver has the higher partial diffusion coefficient in the region of rich aluminium [1]. The interdiffusion coefficients were also calculated in the aluminium-rich phase. The values in table I are in agreement with those of Heumann and Dittrich [1].

The low rate of occurrence of porosity shows that the factors causing porosity are not present or are not sufficiently large [4].

TABLE I Values of \tilde{D} , D_{Al} , D_{Ag}

Specimen	N_{Ag}	\tilde{D} $\text{cm}^2 \text{sec}^{-1}$ $\times 10^{-8}$	D_{Al} $\text{cm}^2 \text{sec}^{-1}$ $\times 10^{-8}$	D_{Ag} $\text{cm}^2 \text{sec}^{-1}$ $\times 10^{-8}$
fig. 2a	0.64	1.73	2.02	1.22
fig. 2b	0.65	9.20	11.41	5.01
fig. 2a	0.04	0.073		
fig. 2a	0.07	0.038		
fig. 2b	0.04	0.180		
fig. 2b	0.07	0.116		

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