JOURNAL OF MATERIALS SCIENCE 13 (1978) 283–290

Mechanical properties characterization of the Pb-Cd eutectic composite

M. SAHOO, R. W. SMITH
Department of Metallurgical Engineering, Queen's University, Kingston, Ontario, Canada K7L 3N6

The tensile and compressive properties of the unidirectionally solidified Pb–Cd eutectic have been examined and compared with those of the Cd–Zn eutectic. In spite of the fact that these two eutectic composites contain volume fractions of almost identical (h c p) reinforcing phase, the former exhibited poor mechanical properties. This has been attributed principally to the relatively lower UTS of the Cd lamellae and to the lower shear strength of the Pb-rich matrix. Off-eutectic alloys showing a coupled eutectic microstructure did not show any appreciable increases in mechanical properties even when grown at lower growth rates where a regular lamellar structure was obtained.

1. Introduction

Some of the areas where unidirectionally solidified eutectic alloys could find potential applications have already been described [1]. However, before any eutectic alloy can be put into commercial use the manner in which its mechanical properties are affected by the various structural parameters must be established. To this end, we have achieved some success in producing a morphological characterization scheme of binary eutectics based on the entropy of solution, ΔS_α, and the volume fraction, V_f, of the minor phase and the growth conditions [2]. In an attempt to characterize their mechanical properties, we have also examined the tensile and compressive properties of a number of anomalous and regular eutectics [3–6] where V_f varied between 6 and 18%.

As part of a continuing programme to characterize the mechanical properties of eutectics, this paper examines the structure and mechanical properties of the unidirectionally solidified Pb–Cd eutectic in comparison with those of the Cd–Zn eutectic whose mechanical behaviour has already been examined in detail [6]. The Pb–Cd eutectic system was chosen for a number of reasons. First of all, both the Pb–Cd and the Cd–Zn eutectics belong to the same morphological group in the eutectic characterization scheme proposed earlier [2]. Secondly, the lamellar reinforcing phases in these two eutectics have an h c p structure and occupy 18% by volume, of their respective eutectics. However, the matrices are respectively f c c and h c p and so are expected to supply different matrix constraints during deformation. In addition, the reinforcing potential of the two h c p phases can be compared to estimate the maximum strength level attainable in each eutectic composite. To this end, the mechanical properties of off-eutectic Pb–Cd alloys, exhibiting a coupled eutectic microstructure, have also been examined in tension and compression.

2. Experimental

Eutectic and off-eutectic Pb–Cd alloys were prepared from 99.9999% pure Pb and 99.999% pure Cd. The experimental procedures for alloy preparation, directional solidification, metallographic examination, tensile and compressive testing were similar to those described in Reference [6] for the Cd–Zn eutectic. However, a furnace temperature of 600° C coupled with a lower eutectic temperature relative to the Cd–Zn eutectic produced a temperature gradient of approximately 10° C mm^{-1} in the liquid ahead of the interface.

The etching reagent consisted of a warm solution of conc. nitric acid and water in the proportion of 1 : 5. Since polishing artifacts readily occur in this system, great care was exercised to ensure that the microstructures prepared were truly representative of those present.

3. Results

3.1. Microstructure

The microstructural form produced in the Pb–Cd eutectic were similar to those produced in the Cd–Zn eutectic. A lamellar microstructure was produced at low growth rates. Typical microstructures for the longitudinal and transverse sections are shown in Figs. 1a and b. The lamellar microstructure was replaced by a cellular one at growth rates greater than 400 mm h^{-1} (Figs. 2a and b)).

It is well known that in several eutectic systems in which a lamellar structure is produced, the interlamellar spacing, λ, measured on the transverse sections should vary with the growth rate, R, according to the equation:

$$\lambda^2 R = \text{constant} \tag{1}$$

This relationship has already been varified in the Pb–Cd eutectic [7–9]. The λ values obtained in the present work have been found to fit into the

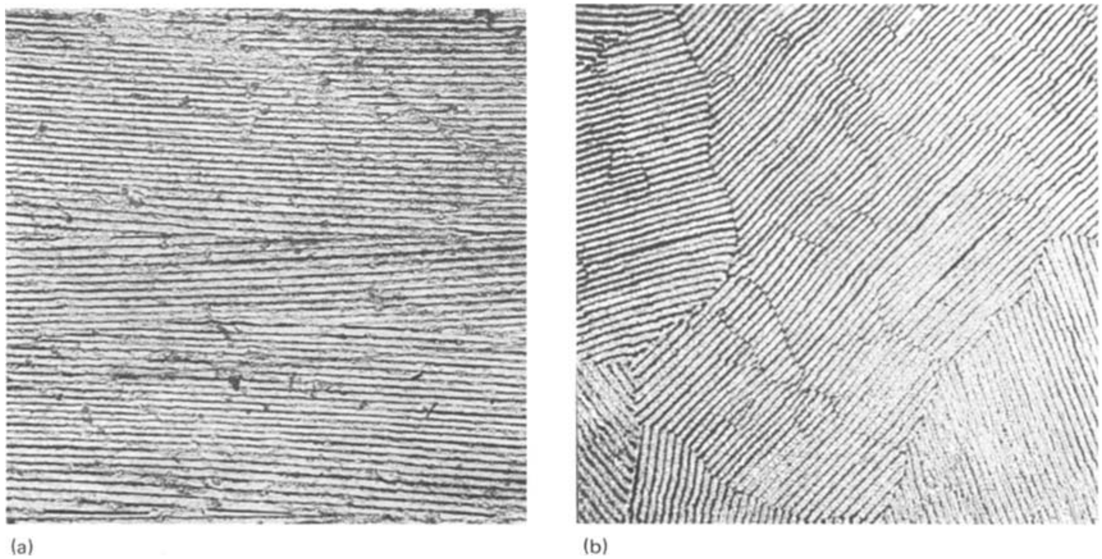
(a) (b)

Figure 1 Optical micrograph of the directionally solidified Pb–Cd eutectic showing the lamellar structure. Growth rate = 12 mm h^{-1} (×400). (a) longitudinal, (b) transverse.

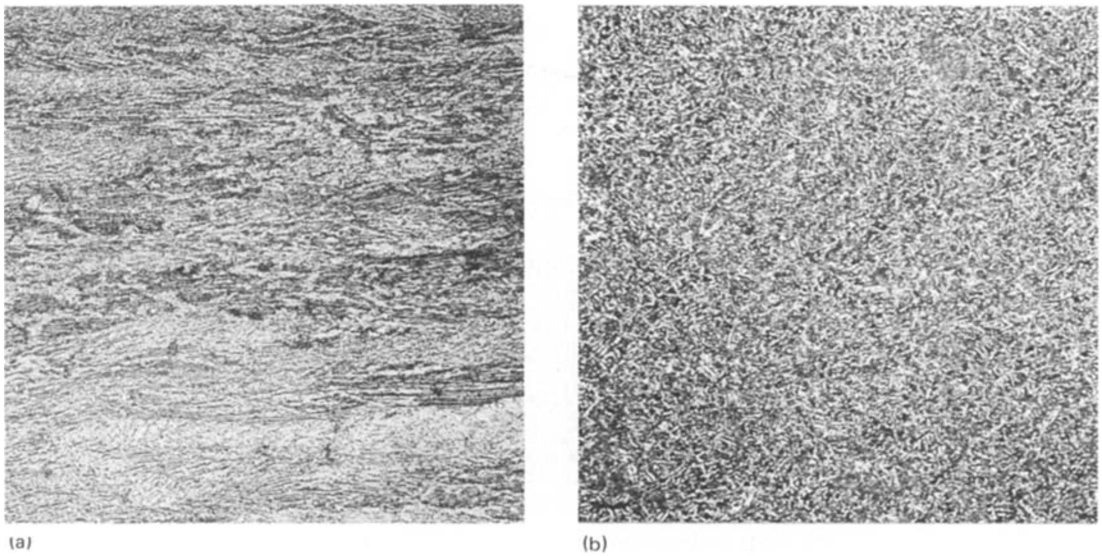
(a) (b)

Figure 2 Optical micrograph of the directionally solidified Pb–Cd eutectic showing the cellular structure. Growth rate = 4000 mm h^{-1} (×400). (a) longitudinal, (b) transverse.

plot of λ versus $R^{-1/2}$ described elsewhere [7–9].

In common with the Cd–Zn system, the high growth rate technique [10, 11] was employed to investigate the range of "coupled" eutectic growth in the Pb–Cd system. Fig. 3 depicts the range of composition as a function of growth rate where a coupled-eutectic microstructure can be obtained.

The composition zone of coupled eutectic growth on the Cd-rich side of the eutectic composition is somewhat larger than that reported by Clark and Elliot [9] who were successful in producing lamellar structures in alloys containing 18 to 25.7 wt % Cd. The temperature gradient in the liquid in their experiments was about 12° C mm^{-1}. In both the Pb–Cd and Cd–Zn systems the eutectic composition corresponds to 17.4 wt % of their respective minor components [12–14]. It is thus evident from Fig. 3 and from Reference 6 that there is an increase in the composition range of composite structures in the Pb–Cd system for identical growth velocities. This difference in composition range may be attributable to the relatively higher temperature gradient employed with this system.

3.2. Mechanical properties

The tensile and compressive properties are summarized in Table I. Typical room temperature stress–strain curves are shown in Fig. 4. These curves are drawn only up to the UTS of the eutectic since failure did not occur at the UTS. The total strain to failure was greater than 7%. By analogy with the Cd–Zn eutectic, the Pb–Cd eutectic can be classified as a ductile composite.

The effect of solidification rate on the tensile and compressive yield strength is shown in Fig. 5. For comparison purposes the tensile and compressive strength data for the Cd–Zn system have been included in Fig. 5. It is evident from Table I and Fig. 5 that both the tensile and compressive strengths increased with growth rate up to a growth velocity of 1200 mm h^{-1}. Further increases

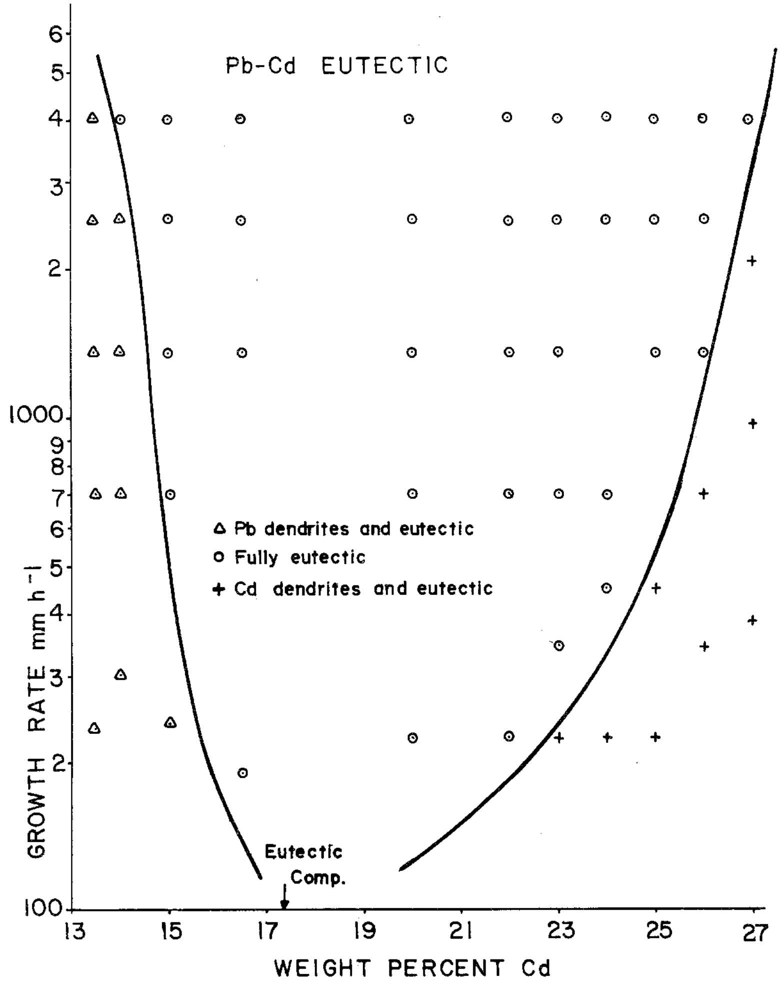

Figure 3 Composition – growth rate plot showing the range of coupled-eutectic microstructure.

TABLE I Room temperature mechanical properties of directionally solidified Pb–Cd eutectics

Growth rate (mm h^{-1})	Structural type*	Tension			Compression		
		0.2% off-set yield strength (N mm^{-2})	UTS (N mm^{-2})	Uniform elongation (%)	0.2% off-set yield strength (N mm^{-2})	UCS (N mm^{-2})	Uniform contraction (%)
As-cast		25.6	36.2	2.1	–	–	–
2.5	L	56.6	58.6	0.6	54.1	62.0	4.0
					58.4	65.0	4.0
12.0	L	53.6	55.2	2.7	–	–	–
		51.6	56.5	3.3	60.7	62.7	1.0
		52.0	55.2	2.2	52.2	58.0	2.5
		53.0	55.2	0.6	64.8	66.0	1.5
155	L	69.4	72.5	2.2	80.0	85.0	3.0
		62.0	69.0	2.6	82.8	89.0	4.5
		62.6	67.7	1.2	81.5	89.7	4.6
400	L	70.2	92.0	1.0	–	–	–
1200	C	84.8	110.0	2.6	80.0	86.3	1.2
		67.6	102.5	3.1	70.4	81.5	3.1
		83.5	110.0	2.3	113.0	136.5	1.4
		75.8	101.0	2.0	110.0	116.0	0.9
1600	C	75.5	100.0	2.9	–	–	–
2800	C	48.2	81.7	3.5	–	–	–
4000	C	39.3	64.6	3.5	69.0	89.7	2.8
					65.0	83.5	3.8

*L = Lamellar, C = Cellular

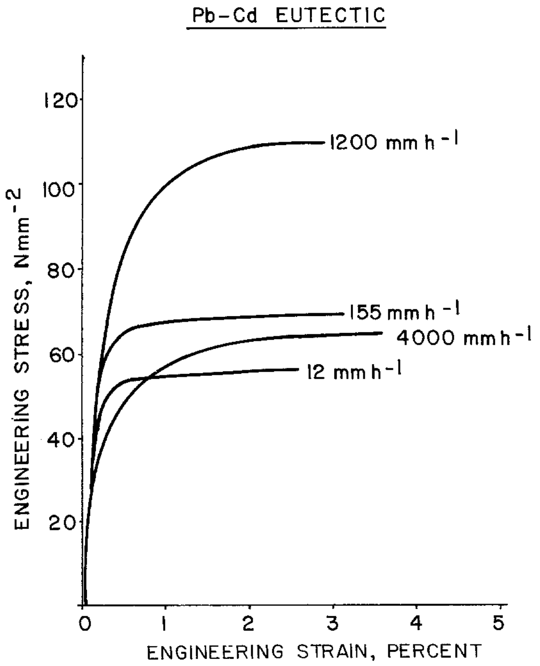

Figure 4 Room temperature engineering tensile stress-strain curves for Pb–Cd eutectic alloys directionally solidified at various rates.

in growth rate produced a significant drop in the strength values. It may also be noted in Table I that for each growth condition the compressive strengths are only slightly higher than the tensile strengths. By contrast, the strength of the Cd–Zn eutectic increased monotonically with the growth velocity up to a growth rate of 4000 mm h^{-1} and its compressive strengths were always higher than the tensile strengths.

Another notable difference lies in the strength levels of these two eutectic composites. The UTS of each eutectic composite has been almost doubled by increasing the solidification rate during unidirectional solidification. However, for any growth condition the Cd–Zn eutectic exhibited higher strength values.

Following the analysis presented in Reference 6, it was found that the tensile and compressive yield strengths of the Pb–Cd eutectic could be related to the growth velocity (up to a growth rate of 1200 mm h^{-1}) by the following expressions:

$$\sigma_{\text{tension}} = 35 + 6.6\,R^{0.21} \tag{2}$$

$$\sigma_{\text{comp}} = 40 + 8.0\,R^{0.22} \tag{3}$$

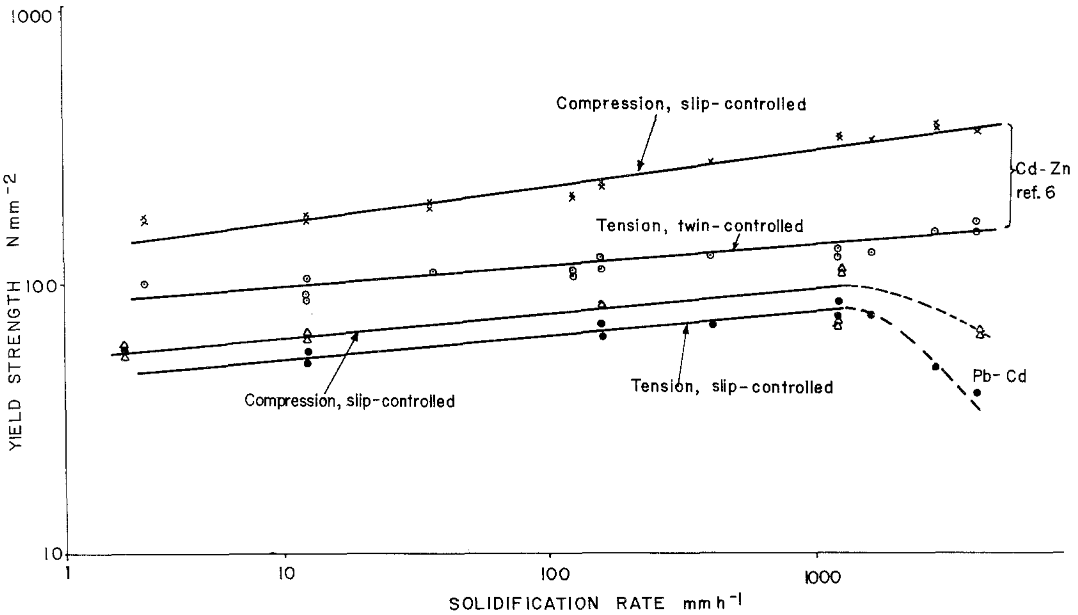

Figure 5 Effect of solidification rate on the tensile and compressive yield strength (0.2% off-set) of the directionally solidified Pb–Cd and Cd–Zn eutectic composites.

Table II summarizes the tensile and compressive strength data for the off-eutectic compositions. It is evident when the mechanical properties of an off-eutectic alloy are compared with those of the eutectic alloy, that, in common with the Cd–Zn system, variations in the volume fraction of the reinforcing phase do not produce any appreciable changes in the mechanical properties for similar growth conditions. This is particularly surprising for the off-eutectic alloys produced at lower growth rates and possessing a regular lamellar structure. Here the volume fraction of the stiffer Cd-

TABLE II Room temperature mechanical properties of directionally solidified off-eutectic Pb–Cd alloys

Wt%Cd	V_f	Growth rate (mm h^{-1})	Micro-structure*	Tension			Compression		
				0.2% off-set yield strength (N mm^{-2})	UTS (N mm^{-2})	Uniform elongation (%)	0.2% off-set yield strength (N mm^{-2})	UCS (N mm^{-2})	Uniform contraction (%)
15	0.153	4 000.0	C	60.8	94.2	4.0	94.0	112.5	1.9
							96.0	113.2	1.9
16.5	0.172	2 800.0	C	65.1	103.4	3.0	113.2	127.6	1.8
							114.5	128.4	2.5
22	0.24	34.0	L	47.3	53.4	5.0	–	–	–
		155.0	L	50.7	67.2	3.5	–	–	–
		1 200.0	C	80.0	101.3	3.0	–	–	–
		2 800.0	C	55.2	89.0	3.2	114.5	122.0	1.0
							114.5	123.0	1.9
23	0.252	1 200.0	C	79.2	99.2	2.6	–	–	–
		4 000.0	C	35.9	56.2	3.3	75.1	85.7	3.10
							71.8	84.4	3.0
25.7	0.285	430.0	L and D	50.7	59.3	4.5	–	–	–
26	0.288	4 000.0	C	57.1	75.1	3.2	91.8	104.0	2.5
							84.0	93.8	2.7

*L = lamellar, C = cellular, D = dendritic rich phase changes from 18% at the eutectic composition to 24% at 22 wt% Cd and so an increase in strength might reasonably be expected.

Barclay and Winegard [15] have examined the tensile properties of the thermally coarsened Pb–Cd eutectic directionally solidified at a rate of 75 mm h^{-1}. Their alloys were prepared from 99.999% pure starting materials and zone-levelled to the exact eutectic composition. The UTS of the as-grown eutectic was only 20 N mm^{-2} for a strain rate of $7.26 \times 10^{-6} sec^{-1}$. This strength value is significantly lower in comparison with the present data where a strain rate of $8.3 \times 10^{-5} sec^{-1}$ was used for tensile testing and from Fig. 5, the UTS of the eutectic is expected to be about 60 N mm^{-2} for a growth rate of 75 mm h^{-1}. In view of this, the effect of strain rate on tensile properties was examined by performing tensile tests at strain rates of 3.3×10^{-5} and $3.3 \times 10^{-4} sec^{-1}$ for specimens directionally solidified at a rate of 155 mm h^{-1}. The average UTS values of the eutectic at these two strain rates were respectively 65 and 74 N mm^{-2}. A sample plot of UTS versus strain rate suggests that the UTS of the eutectic for a strain rate of $7.26 \times 10^{-6} sec^{-1}$ would be 58 N mm^{-2}, again almost three times the value measured by Barclay and Winegard.

4. Discussion

The above results show that the Pb–Cd eutectic composite is relatively weaker than the Cd–Zn eutectic composite. To explain this it is necessary to take account of the various parameters which contribute to the strength of the unidirectionally solidified eutectic composites. Bibring [16] considers these to be concerned with certain characteristics of both the reinforcing phase and the matrix and their rheological interaction at their common interface.

In these two eutectic composites, the volume fraction, ductility, morphology and crystal structure of the reinforcing phases and the interlamellar spacings are similar. The aspect ratio in each is greater than the critical value. In addition the interfacial bonding in lamellar eutectic composites is expected to be strong. This was confirmed by examining the longitudinal sections through fracture surfaces of the two lamellar eutectics which showed broken lamellae close to the fracture surface (See Fig. 8a, Reference 6). Thus the factors necessary to be considered are:

(1) reinforcing phase	(i) inherent strength (ii) perfection
(2) matrix	(i) inherent strength (ii) work hardening capacity and deformability
(3) interface	(i) thermal expansion mismatch (ii) solidification mismatch (iii) contraction effects due to the difference in the Poisson's ratio of the two phases.

In general for maximum reinforcement the reinforcing phase should be stronger and stiffer than the matrix and so, since the strength of vapour-grown Cd and Zn whiskers in the *c*-direction has been reported as 260 and 434 N mm^{-2} respectively [17], the reinforcing potential of Zn lamellae in the Cd-rich matrix is expected to be better than that of the Cd lamellae in the Pb-rich matrix. In addition, since the shear stress of the Cd-rich matrix is higher than that of the Pb-rich matrix when the basal plane of the former is oriented parallel to the tensile or compressive axis [18, 19], its stress bearing capacity will be higher. Thus the bulk properties of the major and minor phases would suggest that the Cd–Zn eutectic should be stronger in general than Pb–Cd.

The extent to which the interphase boundaries in each eutectic significantly influence mechanical behaviour is not clear. To increase the UTS of any eutectic, the work hardening capacity of its ductile matrix must be increased. Work hardening occurs due to interaction of slip dislocations with the existing dislocation network at the strengthening phase/matrix interface and so will depend on the deformation modes of the matrix and the cyrstallographic nature of the interphase boundary.

In the case of the Pb–Cd eutectic, the Pb-rich matrix is f c c and hence plastic deformation can take place by the operation of many slip systems to cause work hardening of the matrix. In addition, since the difference in the Poisson's ratios of the Pb and Cd phases ($\nu_{Pb} = 0.44$, $\nu_{Cd} = 0.30$) is large [20, 21], some matrix hardening is induced. By contrast, the Cd-rich matrix of the Cd–Zn eutectic has an h c p structure and so its work hardening capacity is expected to be lower than that of the Pb-rich matrix because of the opeation of limited slip systems. This leads to a twinning mode of deformation in tension [6] whereas in the case of the Pb–Cd eutectic, deformation in both tension and compression takes place by slip.

As to the crystallographic nature of the interphase boundaries in each eutectic, little strain is expected to be introduced into Pb–Cd during cooling from the melt since the linear expansion coefficients (α) of each phase is similar ($\alpha_{Pb} = 29.1 \times 10^{-6°} C^{-1}$ and $\alpha_{Cd} = 31.8 \times 10^{-6°} C^{-1}$, [20]), and they would be expected to assume a low mismatch interface. (No crystallographic data has been reported for the Pb–Cd eutectic). In Cd–Zn, on the other hand, the large difference in the expression coefficients of the two phases [6] is likely to promote considerable elastic and plastic strain on cooling so that, in spite of similar Poisson's ratios ($\nu_{Zn} = 0.33$, [21]) and almost perfect epitaxy across the common (0001) boundary plane, a large dislocation density is likely to exist at the interphase boundary at the test temperature.

Thus it may be concluded that the crystallographic structure of the interphase boundary probably does contribute significantly to the superior strength of the Cd–Zn eutectic.

The changes in mechanical properties with growth rate, noted in Table I and Fig. 5, are interesting. It is seen that the tensile and compressive properties of the Pb–Cd eutectic composite increase monotonically with the growth rate up to 1200 mm h^{-1} following which a gradual decrease in these properties takes place. These increases are associated with the structural refinement of the eutectic. Some of the possible reasons for such strengthening have been discussed earlier [4–6, 16]. Given this, then the decreases in the mechanical properties at growth rates greater than 1200 mm h^{-1} tend to indicate that considerable misalignment and irregularity of the Cd lamellae and the growth direction resulting from cellular growth causes much of the reinforcing potential of the Cd phase to be lost. To this end, Fig. 2a may be compared with Fig. 6 which is the optical micrograph of the Pb–Cd eutectic directionally solidified at a growth rate of 1200 mm h^{-1}.

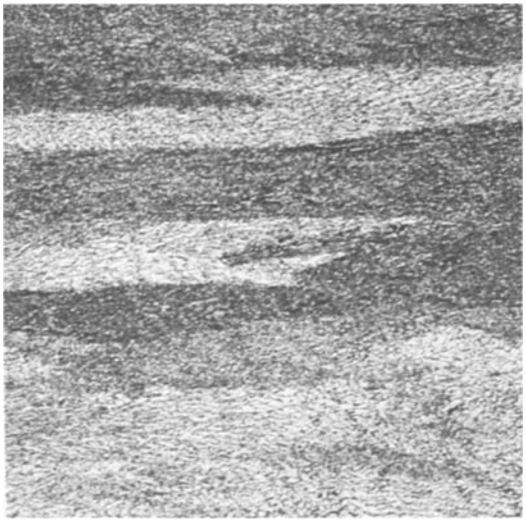

Figure 6 Optical micrograph of the directionally solidified Pb–Cd eutectic showing the cellular structure. Growth rate = 1200 mm h^{-1} (X 500).

Table I shows that the compressive strength of the Pb–Cd eutectic is only slightly higher than the tensile strength for any growth rate. This is in marked contrast to the behaviour reported for other eutectic alloys where dramatic differences in the compressive and tensile strengths were observed [3–6, 22–24]. This latter form of response is usually attributed to the residual stresses due to differences in the thermal expansion coefficients of the two phases and so would not be expected to be present in Pb–Cd.

It was noted from Table II that for similar growth conditions the mechanical properties of the hyper-eutectic Pb–Cd alloys were in no way superior to those of the eutectic composition. In this eutectic system, the factor $\lambda^2 R$ is known to be independent of composition [9]. Thus, at any growth rate the interlamellar spacing is not expected to change due to increases in the Cd content of the alloy to cause added advantages in the mechanical properties. It may also be said in passing that the poor mechanical properties of the off-eutectic alloys at extremely fast growth rates lend support to our previous conclusions about the misalignment and irregularities in the Cd lamellae.

5. Conclusions

The mechanical behaviour of the unidirectionally solidified Pb–Cd eutectic composite, consisting of Cd lamellae in the Pb-rich f c c matrix, is similar to that of the Cd–Zn eutectic composite where Zn lamellae reside in a Cd-rich h c p matrix with the following notable exceptions:

(1) the former is weaker than the latter for any growth condition

(2) the strength of the Pb–Cd eutectic is reduced when grown at a rate in excess of 1200 mm h^{-1}.

(3) no significant differences between the tensile and compressive properties of the Pb–Cd

eutectic could be observed for any growth condition; the strength in compression being slightly higher than in tension.

Acknowledgements

We are indebted to the National Research Council of Canada for financial support.

References

1. M. J. SALKIND, F. D. LEMKEY and F. D. GEORGE, "Wisker Technology," edited by A. F. Levitt (Wiley-Interscience, New York, 1970), p. 343.
2. M. N. CROKER, R. S. FIDLER and R. W. SMITH, *Proc. Roy. Soc.* **A335** (1973) 15.
3. M. SAHOO and R. W. SMITH, *Metal Science* **9** (1975) 217.
4. *Idem, Can. Met. Quart.* **15** (1976) 1.
5. *Idem, J. Mater. Sci.* **11** (1976) 1125.
6. M. SAHOO, R. A. PORTER and R. W. SMITH, *J. Mater. Sci.* **11** (1976) 1680.
7. G. A. CHADWICK, *JIM*, **92** (1963–64) 18.
8. A. MOORE and R. ELLIOT, "The Solidification of Metals," (Iron and Steel Inst., London, 1968) p. 226.
9. J. N. CLARK and R. ELLIOT, *Metal Science* **10** (1976) 101.
10. H. E. CLINE and J. D. LIVINGSTON, *Trans. Met. Soc. AIME* **245** (1969) 1987.
11. R. S. BARCLAY, H. W. KERR and P. NIESSON, *J. Mater. Sci.* **6** (1971) 1168.
12. M. HANSEN and K. ANDERKO, "Constitution of Binary Alloys," (McGraw-Hill, New York, 1958).
13. R. P. ELLIOT, "Constitution of Binary Alloys," First Supplement (McGraw-Hill, New York, 1965).
14. F. A. SHUNK, "Constitution of Binary Alloys," Second Supplement (McGraw-Hill, New York, 1969).
15. R. S. BARCLAY and W. C. WINEGARD, *Mat. Sci. and Eng.* **13** (1974) 291.
16. H. BIBRING, Proceedings of the Conference on *In Situ* Composites, Vol. II, (National Academy of Eng., Washington D.C., 1973) p. 1.
17. R. V. COLEMAN, B. PRICE, and N. CANBERA, *J. App. Phys.* **28** (1957) 1360.
18. F. WEINBERG, *Trans. Met. Soc. AIME* **242** (1968) 2111.
19. N. S. STOLOFF and M. GENSAMER, *ibid* **224** (1962) 732.
20. C. H. SMITHELS, "Metals Reference Book Vol. III," (Butterworths, London 1967).
21. E. SCHMID and W. BOAS, "Plasticity of Crystals," (Chapman and Hall, London, 1968) p. 19.
22. E. R. THOMPSON, D. A. KOSS and J. C. CHESTNUT, *Met. Trans.* **1** (1970) 2807.
23. A. PETTANAIK and A. LAWLEY, *ibid* **2** (1971) 1529.
24. W. R. KRUMMHEUER and H. ALEXAMDER, *J. Mater. Sci.* **9** (1974) 229.

Received 21 March and accepted 8 June 1977.