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High-temperature properties of rapidly solidified Al–Be alloys

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Rapidly solidified Al–Be binary alloys (10%, 20%, and 40% Be by weight) have been produced by melt-spinning techniques. The microstructures have been evaluated and the elevated temperature mechanical properties have been characterized over a range of strain rates. Despite the fact that the materials exhibited duplex microstructures resulting from high-temperature processing, they showed behaviour typical of dispersion strengthened alloys. The mechanical properties at elevated temperature can accurately be described by the equation $\dot{\varepsilon} = A(\sigma/E)^{12} \exp(-126\,\mathrm{kJ\,mol^{-1}}/RT)$, where $\dot{\varepsilon}$ is the deformation rate, σ is the stress, E is the modulus, A is a material constant, and RT has its usual meaning. A direct comparison of the deformation properties was made between binary Al–Be composition and pure aluminium as baseline, as well as between Al–Be and some high temperature aluminium alloys. The Al–Be alloys do not exhibit good high-temperature strength when compared with other high-temperature aluminium alloys, e.g. Al–Fe–Ce alloys. This is a result of particle coarsening and agglomeration during processing and testing.

1. Introduction

Lithium and beryllium are the only two elemental additions to aluminium that can simultaneously increase the modulus and reduce the density. Aluminium–lithium alloys have been extensively studied recently because of their potential applications for aerospace structures [1–3]. In contrast, the Al–Be alloy system has not been thoroughly examined to date. This may be partly because of the difficult problems associated with the production of such compositions. However, the potential weight savings as a result of density reduction, and improvements in the elastic modulus, still make this alloy group worthy of evaluation.

A previous development known as Lockalloy [4], which had a composition of 38% Al–62% Be, was once commercially available; however, this is a beryllium-based alloy. Although aluminium-rich ternary alloys based on Al–Be–Mg have been evaluated by a number of researchers [5], the materials were either prepared by conventional powder metallurgy techniques or prepared by the chill-casting technique. In this paper, Al–Be alloys prepared from melt-spun ribbons are examined and their mechanical properties at elevated temperature are presented.

2. Experimental procedure

Binary alloys of Al–Be were prepared by arc melting high-purity ingot stock aluminium and electrolytic induction-melted beryllium stock. The arc-melted buttons were converted to melt-spun ribbon having the approximate dimensions of 1 mm wide and less than 0.1 mm thick. Melt spinning was carried out using an apparatus and techniques described elsewhere [6]. The alloy ribbons were comminuted, hot pressed at 482° C, and extruded at about 427° C through a 20 : 1 reduction die with a 2 : 1 aspect ratio. The extruded bar was further rolled down to a sheet with a final thickness of 1.9 mm, from which the tensile coupons were machined.

Hardness measurements were carried out using either a conventional Rockwell machine or a Vickers microhardness machine. Usually, five measurements were taken from each specimen. Elevated-temperature tensile tests were conducted with a conventional Instron machine controlled by a HP3497-based data acquisition system. A resistance heating furnace was used to bring the test specimen up to temperature. A preheat of approximately 45 min was used to stabilize the test temperature. All tests were carried out under conditions of constant true strain rate, ranging from 5×10^{-5} to $10^{-3}\,\mathrm{sec}^{-1}$.

Microstructures and fracture surfaces were characterized using either optical metallography or scanning electron microscopy.

3. Results and discussion

Microstructures of as-spun Al–Be ribbons are shown in Fig. 1. The microstructure of Al–10% Be is relatively fine and uniform and is similar to the microstructure observed in as-melt-spun Al–10% Be–3% Li [7]. However, for both Al–20% Be and Al–40% Be alloys the microstructures can be clearly divided into two zones. The side which has a faster cooling rate has a much finer microstructure as compared to the other side. The dual microstructure observed in the materials with high beryllium content is a result of the poor thermal conductivities as

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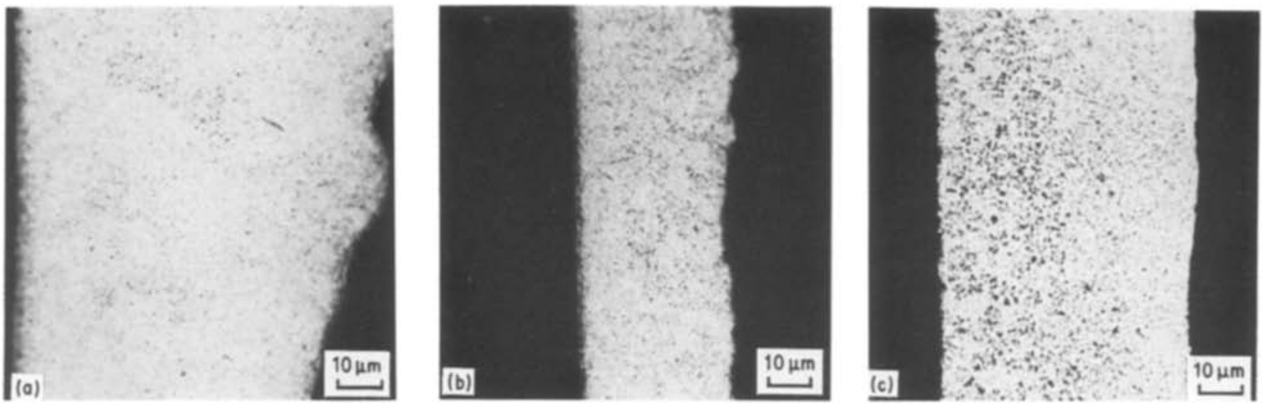

Figure 1 Microstructures of as-spun Al–Be ribbons. Large α-Be particles are occasionally present in the Al–20% Be and Al–40% Be ribbons. (a) Al–10% Be, (b) Al–20% Be, (c) Al–40% Be.

compared to that in the 10% Be material. The contrast of the dual microstructure in the as-melt-spun ribbons becomes more apparent as the beryllium content of the alloy increases. In the base of Al– 40% Be alloy, the beryllium particle size in the slow cooling zone in the ribbon is as coarse as 1 to 2 µm.

Differential scanning calorimetric analyses from the as-melt-spun Al–Be ribbons do not show any exothermic or endothermic activity prior to the melting of the ribbon. This is consistent with the fact that there are no stable intermetallic compounds between aluminium and beryllium. It may further suggest that a metastable Al–Be phase either does not exist or the thermodynamic heat of formation of such a phase is too low to be measured by the equipment used.

The microstructures of the Al–Be alloys after consolidation and extrusion are shown in Fig. 2. It is noticed that all of the compositions show irregular banded structures (coarse and fine) and beryllium particles are heterogeneously distributed. The band width for the coarse structure ranges from 20 to 100 µm. Despite the fine dispersion of beryllium in the fine zone, in the coarse zone the beryllium particle sizes are nonuniform and sometimes as large as 10 µm. In the case of Al–40% Be, because of the very high beryllium content, continuous networks of beryllium form in the coarse zones. As mentioned earlier, the consolidation and extrusion temperatures are 482 and 427° C, respectively. Apparently, the consolidation temperature was sufficiently high to cause the Al–Be to recrystallize partially and coarsen. The recrystallized areas were subsequently elongated during extrusion. This resulted in the duplex microstructure. It should be noted that the size of the coarse zones in the extrudate is not related to the dimensions of the coarse zones observed in the original as-spun ribbon. The microstructural coarsening during consolidation suggests that beryllium particles are not effective grain-growth inhibitors for aluminium alloys. The coarsening is caused by fast diffusion of beryllium in aluminium, as will be discussed later.

The mechanical properties of the two microstructural zones are also different, as expected. For example, the microhardness for Al–10% Be in the fine zone is consistently about HV 66. However, for the coarse zone it ranges from HV 30 to 50, depending on the exact beryllium distribution in that particular zone. For Al–40% Be, because of the higher beryllium content, hardness numbers are much higher and are HV 160 and 80 for the fine and coarse zones, respectively.

Despite the prior high-temperature exposure, the microstructure of the extruded binary alloys still appears to be unstable upon subsequent lower temperature exposure. Fig. 3 shows the room-temperature hardness values for both Al–10% Be and Al–20% Be that have been exposed to various elevated temperatures

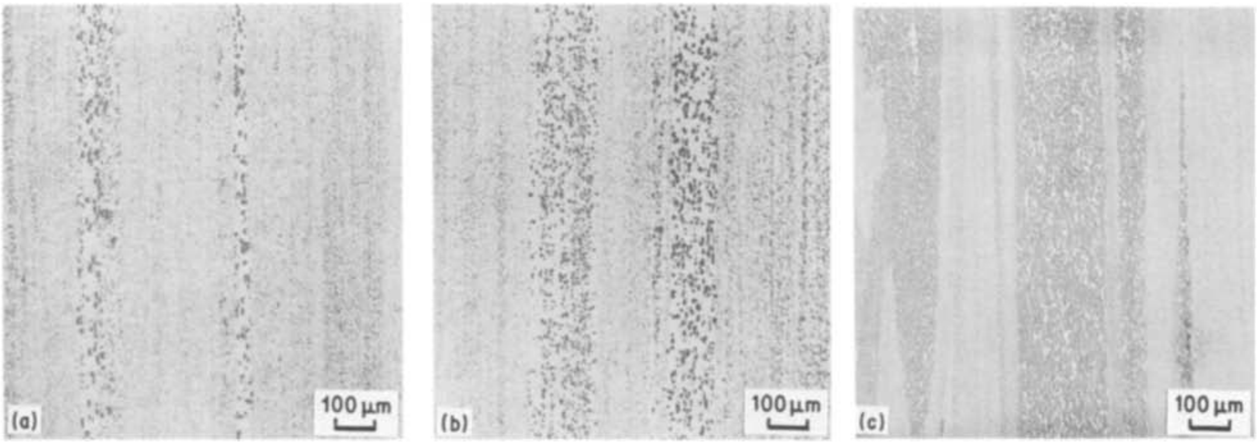

Figure 2 Microstructures of as-extruded Al–Be alloys. The microstructures consist of a banded structure containing coarse and fine zones. The beryllium phase forms a continuous network in the coarse zone in the Al–40% Be alloy. (a) Al–10% Be, (b) Al–20% Be, (c) Al–40% Be.

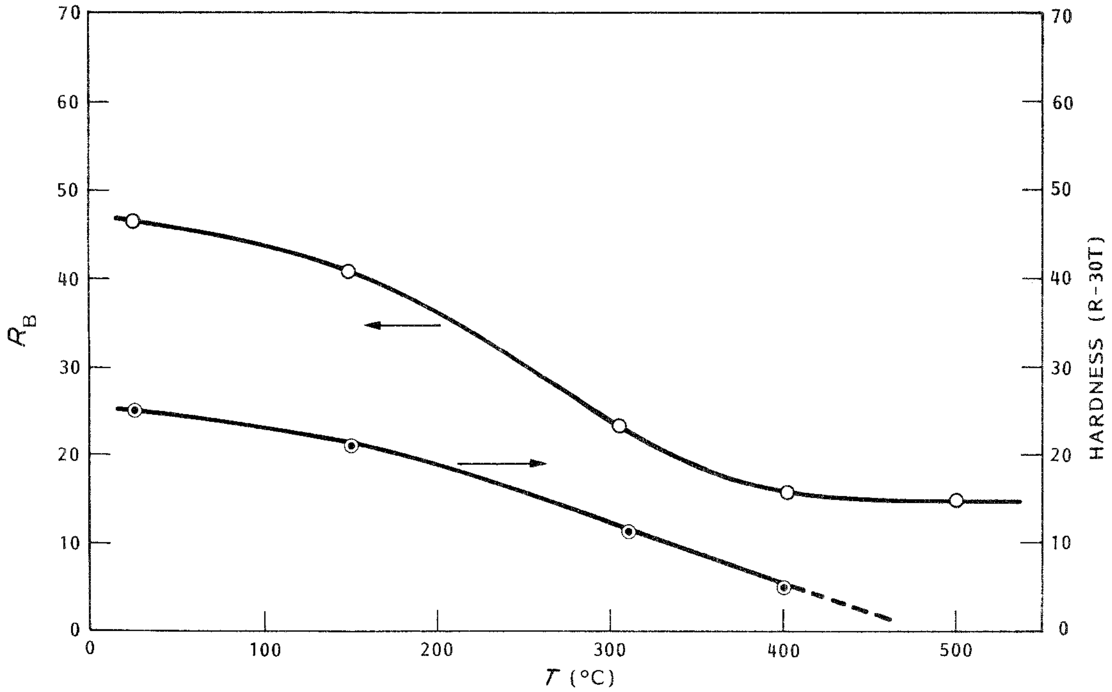

Figure 3 Softening behaviour of (⊙) Al–10% Be and (○) Al–20% Be alloys that have been annealed at various temperatures for 1 h.

for 1 h. It is a general trend that the hardness for both materials decreases continuously as the exposure temperature increases. The relative amount of reduction in hardness, especially for the Al–10% Be alloy, is difficult to account for simply by assuming that the softening can be attributed to dislocation annihilation during the thermal annealing. The softening behaviour is more likely may be a direct result of microstructural coarsening; in particular, the coarsening of beryllium particles at elevated temperatures. This is consistent with the fact that the relative amount of reduction in hardness is less for materials with a high beryllium content. In fact, for Al–40% Be, the hardness values for as-extruded and as-extruded and then thermally annealed materials are almost identical. For a material that has a higher beryllium content, the beryllium particle size is always coarser than for the lower beryllium content. The particle coarsening rate is expected to be slower in such a material by comparison with an alloy that initially contains a finer particle dispersion. Because of the duplex nature of the microstructure, it is very difficult to measure directly the coarsening rate of the beryllium dispersoids. However, microhardness measurements from an Al–10% Be sample show that the hardness values in the fine zone reduce from HV 67 for the as-extruded condition to HV 61 for the material that has been further exposed at 232° C for 30 min.

Although the room-temperature mechanical properties for Al–Be alloys have been characterized elsewhere [8], their tensile properties at elevated temperatures have not been reported. In this study, tensile tests from 121 to 232° C were carried out for both Al–10% Be and Al–20% Be alloys. Both alloys are ductile and have elongations to failure, typically, of 40 to 50%. The tensile stress–strain curves normally exhibit a steady-state type behaviour near the end of the tensile test. In general, the flow stress increases with the strain rate and decreases with the temperature. Also, the Al–20% Be alloy is stronger than Al–10% Be under the same test conditions. A log–log plot of the strain rate against flow stress is shown in Fig. 4 for Al–20% Be. As demonstrated in the figure, the deformation behaviour of the material can be described by the equation $\dot{\varepsilon} = k\sigma^n$ where $\dot{\varepsilon}$ is the true strain rate, σ is the flow stress and k is a constant; the stress exponent, n, is approximately equal to 12. Although it is not shown in the figure, the value of n for Al–10% Be is also a constant and approximately equal to 12. For pure aluminium, the value of n is between 4 and 5 [9, 10]. A high stress exponent is a typical characteristic of dispersion-strengthened alloys [11, 12] and some metal matrix composites [13]. This stress–strain rate relationship in Al–Be is to be expected because beryllium is virtually insoluble in solid aluminium and the beryllium particles act as dispersoids in the aluminium matrix. In addition, the beryllium dispersoids are essentially non-deformable under the test conditions in this study.

When the Al–Be alloys are subjected to mechanical loading along the longitudinal direction, i.e. the direction which exhibits the elongated grain structure, their duplex microstructures are oriented so that both banded structures are under isostrain conditions. In this case, the fine-zone area, which exhibits a higher mechanical strength than that of the coarse-zone area, will dominate the deformation behaviour of the material. This is particularly true in the present case, because the volume ratio for the coarse-zone volume fraction is estimated to be less than 20%.

The activation energy for the high-temperature deformation for both Al–10% Be and Al–20% Be was calculated to be approximately 126 kJ mol^{-1}, which is comparable to 138 kJ mol^{-1}, the activation energy for self-diffusion of aluminium [14, 15]. This result suggests that the deformation process is probably controlled by a dislocation-climb-like mechanism [16]. Similar results have been reported in a number of other materials [17].

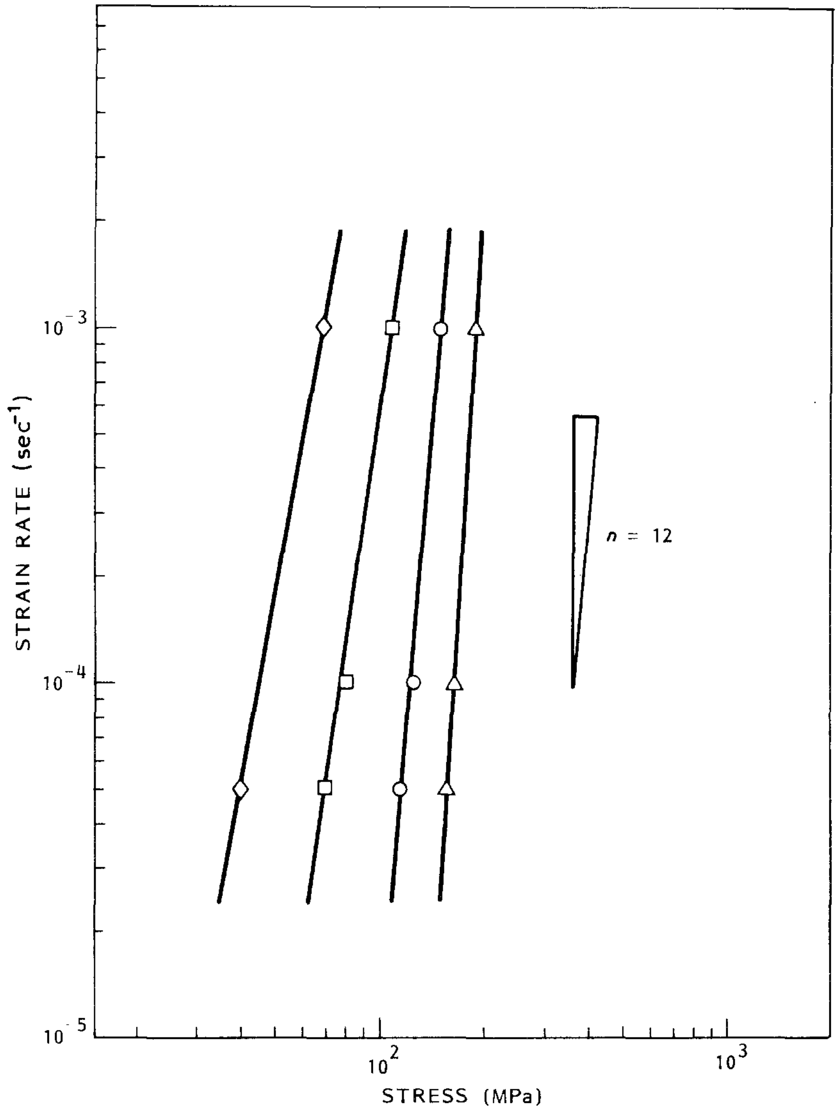

Figure 4 Strain rate as a function of flow stress for Al–20% Be at elevated temperatures. The stress exponent, *n*, is approximately equal to 12. (△) 121° C, (○) 177° C, (□) 232° C, (◇) 288° C.

In summary, the deformation behaviour of Al–Be alloys at elevated temperature can be well described by the equation

$$\dot{\varepsilon} = A\,(\sigma/E)^{12}\exp\,(-Q/RT)$$

where $\dot{\varepsilon}$ is the deformation rate, σ is the flow stress, E is the modulus, Q is the activation energy for the deformation, and A is a material constant. The values of A are $2.5 \times 10^{50}\,\text{sec}^{-1}$ and $7 \times 10^{47}\,\text{sec}^{-1}$ for the Al–10% Be and Al–20% Be alloys, respectively. A summary of the normalized data plotted as log $(\dot{\varepsilon}/D)$ against log (σ/E) is shown in Fig. 5. The correlation between the data and the equation is reasonably good. In order to make a direct comparison, the creep data for pure aluminium [9, 10] is also displayed in the same figure. Although Al–Be alloys are much stronger than aluminium. It is misleading to consider Al–Be alloys as high-temperature alloys from the data of this figure. For example, the high-temperature Al–8% Fe–4% Ce alloy exhibits a strength of 350 MPa at 230° C. For the Al–20% Be alloy the strength is only about 110 MPa under the same conditions [18]. Even on a density-compensated basis, the strength for Al–Be is still far too low to compete with other high-temperature aluminium alloys [18].

The fracture surfaces of tensile-tested samples in general show dimpled-type features indicating ductile fracture, such as the examples shown in Fig. 6. This is consistent with the result that all test samples exhibited good tensile elongations-to-failure. It is interesting to note that the fracture surface also shows a duplex nature in dimple spacing, and can be related to the fine and coarse zones seen in the optical microstructure, as discussed earlier. The average dimple spacing for the fine zone is about 3 μm, while it is nearly 8 μm for the coarse zone. It is also noted that fracture of the samples almost invariably originated from a coarse zone or a coarse beryllium particle.

4. Conclusions

Rapidly solidified Al–Be binary alloys have been produced. A duplex microstructure has been observed in the alloys which was due to fast coarsening of beryllium particles during the consolidiation and extrusion processes. The mechanical behaviour of the alloys has also been characterized at temperatures between 121 and 232° C. The stress exponent was measured to be about 12 and the apparent activation energy for the tensile deformation was calculated to be approximately 126 kJ mol^{-1}, which is close to the activation energy for self-diffusion in aluminium. In general, the alloys behave like a conventional dispersion-strengthened alloy. Although binary Al–Be alloys appear to be much stronger than aluminium

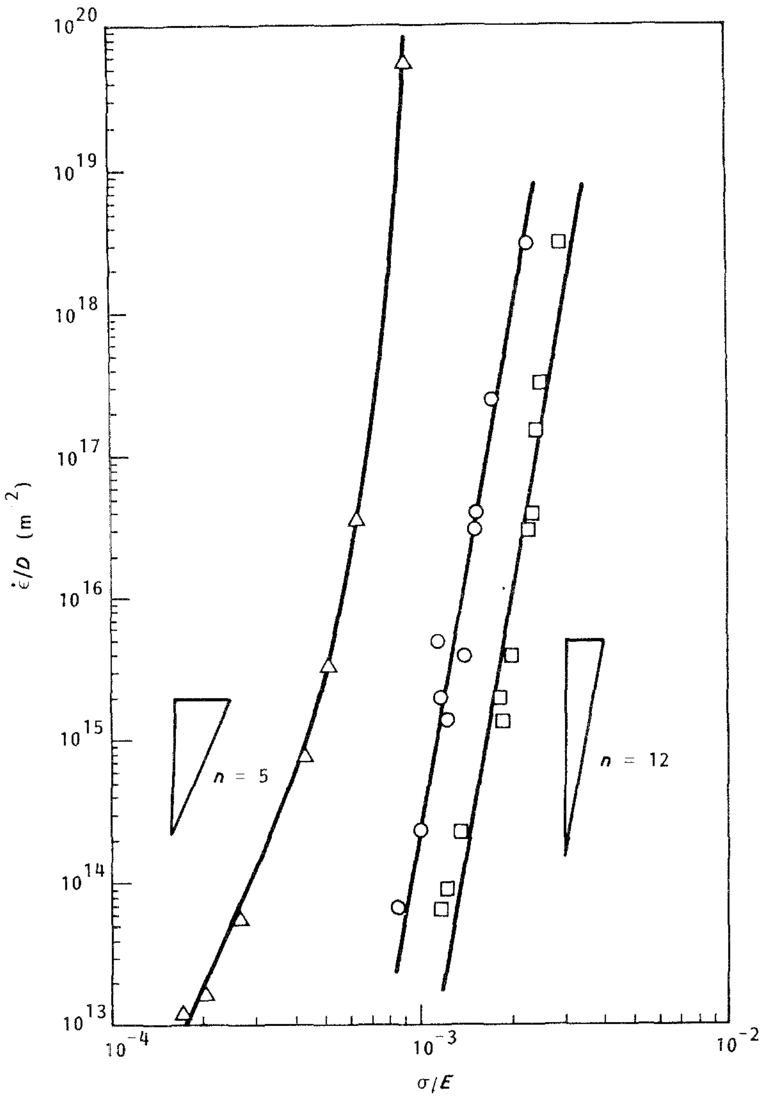

Figure 5 Log ($\dot{\varepsilon}/D$) plotted against log (σ/E) for (○) Al–10% Be, (□) Al–20% Be. The data for pure aluminium (△) [9, 10] are also included in the figure for a direct comparison. Both Al–Be alloys are much stronger than pure aluminium. The high stress exponent at a high stress regime for pure aluminium represents power-law breakdown.

itself, these alloys do not exhibit high-temperature strength retention by comparison with high-temperature aluminium alloys, e.g. Al–Fe–Ce. This is primarily because of the relatively rapid diffusion of beryllium in the aluminium matrix at elevated temperature which leads to coarsening of the beryllium phase.

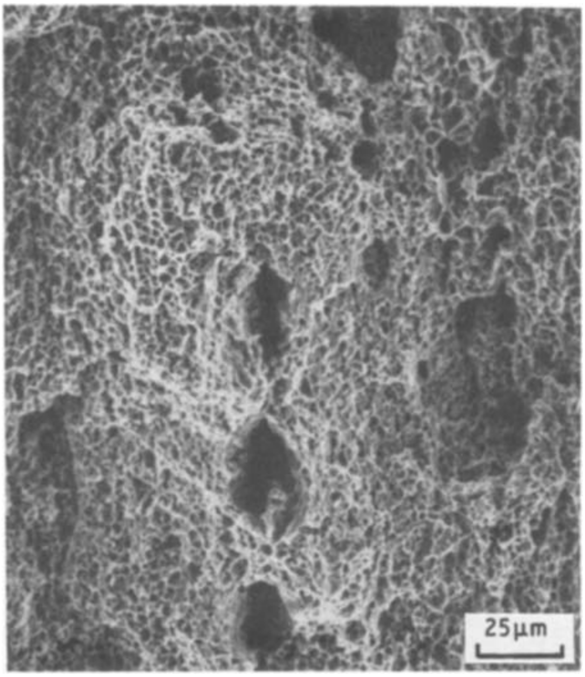

Figure 6 Typical dimpled, ductile fracture in Al–Be alloys.

Acknowledgements

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