JOURNAL OF MATERIALS SCIENCE 27 (1992) 5931–5934

Plastic behaviour of $CuAl_2$

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The plastic behaviour of $CuAl_2$ was studied by compression testing of single crystals and polycrystals in the temperature range 300–575 °C. While single crystals were grown from the melt by the Bridgeman technique, ingot and powder metallurgy routes were adopted for polycrystalline specimens. In addition to exploring their flow behaviour, the deformation mechanism was assessed through thermal activation analysis. It was observed that $CuAl_2$ failed in a brittle manner in compression below 375 °C and its ductility improved progressively with temperature. The brittle–ductile transition (BDT) temperature was influenced by the initial dislocation density but not by the grain size. The strong temperature dependence of flow stress and grain size strengthening effect as per the Hall–Petch relation, were dominant up to nearly the melting temperature of $CuAl_2$. The measured activation parameters for deformation suggest that the Peierls mechanism is rate controlling over the investigated temperature range.

1. Introduction

The mechanical behaviour of intermetallics is of considerable interest in exploiting their potential for structural applications at elevated temperatures. Intermetallics in general have good phase stability, corrosion resistance and high strength. However, their structural use has been restricted because of their poor ductility. An understanding of factors affecting their ductility would be beneficial in this regard. The brittleness of intermetallics has been explained in terms of several factors [1], such as the limited number of slip systems, restricted cross-slip, a high Peierls stress associated with a large slip vector [2], and difficulty in transmitting slip across grain boundaries. Attempts [3] towards overcoming their brittleness have been in terms of micro- and macro-alloying, microstructural control, fibre strengthening and toughening. Some intermetallics of commercial interest are TiAl, Ti_3Al, Ni_3Al, $(Fe, Co, Ni)_3V$, Zr_3Al and B2 aluminides such as NiAl, FeAl and CoAl. They can be processed by ingot metallurgy and powder metallurgy techniques. In some cases, hot rolling and forging are possible because of their superplastic behaviour. Some are now being tested under service conditions in the form of turbine blades and other components [4]. An advantage in some intermetallics is their anomalous increase in strength with temperature.

This study concerns the mechanical behaviour of $CuAl_2$ which is of interest both from the fundamental viewpoint of one class of intermetallic compounds and as a reinforcing phase in the metal matrix composites. Previous work on this compound is briefly reviewed here. The elastic constants of $CuAl_2$ were measured by ultrasonic wave velocity and extensometric techniques [5] and the values of Young's modulus and Poisson's ratio thus measured are 106 GPa and 0.31, respectively. The operative slip systems in single crystals of $CuAl_2$ were studied by Kirsten [6], and Ignat and Durand [7]. The hot hardness technique was employed by Petty [8] to investigate the plastic behaviour of $CuAl_2$. Dey and Tyson [9] reported a detailed study on the plasticity of polycrystalline $CuAl_2$ in tension and compression at various temperatures. They observed that $CuAl_2$ exhibits some ductility in tension at temperatures above 450 °C, whereas the brittle to ductile transition (BDT) temperature in compression is about 400 °C. Through measurement of the activation parameters, they identified the rate-controlling mechanism of deformation as the Peierls stress in the lower temperature regime. The present study was aimed at further exploring the flow behaviour of $CuAl_2$. The factors considered were the brittle to ductile transition in single crystals compared to polycrystals, the effect of the nature of the grain boundaries and the initial dislocation density, and the operative dislocation mechanism over the whole temperature range.

2. Experimental procedure

The alloy of nominal composition 53.5 wt % Cu and 46.5 wt % Al was prepared by induction melting in a split graphite crucible under an argon atmosphere starting with 99.9% purity metals. Subsequently the alloy was remelted in another graphite crucible to obtain rods of 7 mm diameter. Some of these rods were remelted and grown into single crystals by the Bridgeman technique. In order to detect grain boundaries, if present, these rods were macroetched with $FeCl_3$ solution, but the orientations of the single

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crystals were not characterized. In addition to the specimens of single crystals and polycrystals (as-cast), the powder metallurgy (P/M) route was employed to prepare another set of polycrystalline specimens. For this purpose, the powder prepared by ball milling was hot pressed at 540 °C into cylindrical specimens 7 mm diameter and 8 mm long. The density of these specimens was better than 95% theoretical. Compression test specimens, 7 mm diameter and 8 mm long, were also prepared from as-cast and single-crystal rods making use of a diamond wheel. Careful polishing of the end faces of each compression specimen was carried out to ensure smooth and parallel surfaces.

All the mechanical tests were performed in compression in the temperature range 300–575 °C on an MTS servohydraulic machine. In addition to isothermal tests at a given crosshead speed, tests involving changes in test temperature and crosshead speed were carried out to assess thermally activated flow parameters. Standard procedures of specimen preparation followed by electrolytic polishing and etching with a solution of HNO_3 and CH_3OH in 1:2 proportion were employed for metallographic examination.

3. Results and discussion

3.1. Flow behaviour

As mentioned in Section 2, different processing routes were adopted to obtain wide variations in grain size. Single crystals, specimens in the as-cast condition with a grain size of 160 μm, and P/M processed specimens with a grain size of 10 μm, were tested over a wide temperature range. All specimens failed in a brittle manner at 375 °C and below. Above 375 °C, their ductility progressively increased with temperature. Their plastic flow is accompanied by strain hardening and/or strain softening as seen from typical flow curves presented in Fig. 1. The P/M specimens were observed to exhibit lesser ductility than single crystals at all temperatures. Furthermore, the P/M specimens fractured in shear mode following deformation of varying amounts depending on the test temperature.

In order to study the effect of the nature of grain boundaries, some of the specimens processed by the ingot metallurgy route were prestrained to 10% strain at 575 °C. Metallographic examination of the prestrained specimens indicated recrystallized microstructure without a significant change in grain size. The flow curves of the recrystallized specimens are compared with those of the as-cast condition in Fig. 2. It may be noted that at 420 °C the recrystallized specimen deformed extensively, whereas the as-cast specimen failed in a brittle manner.

The effect of initial dislocation density on the flow behaviour and brittle to ductile transition was investigated in the following manner. Some of the single-crystal specimens and those of the as-cast condition were deformed to a 3% plastic strain at 475 °C. Subsequently they were tested at lower temperatures and their flow behaviour was compared to those without prestrain. The above prestrain did not result in recrystallization as observed by metallography. Thus the effect of prestrain can be linked to an increase in the initial dislocation density. From the flow curves shown in Figs 3 and 4, an improvement in ductility is

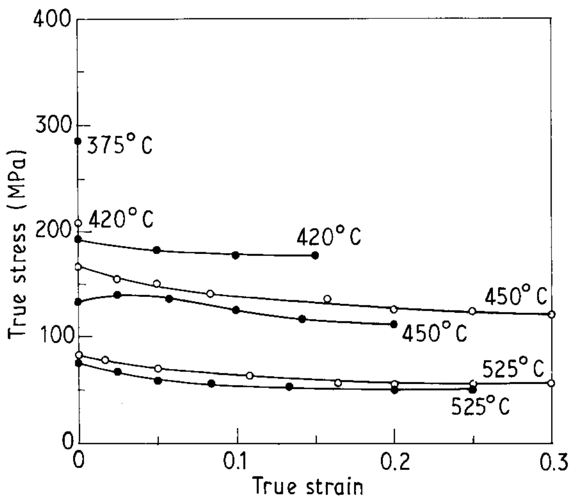

Figure 2 Stress–strain curves of (○) as-cast and (●) recrystallized specimens (cast and prestrained 10% at 575 °C). $\dot{\varepsilon} = 5 \times 10^{-5}\ s^{-1}$.

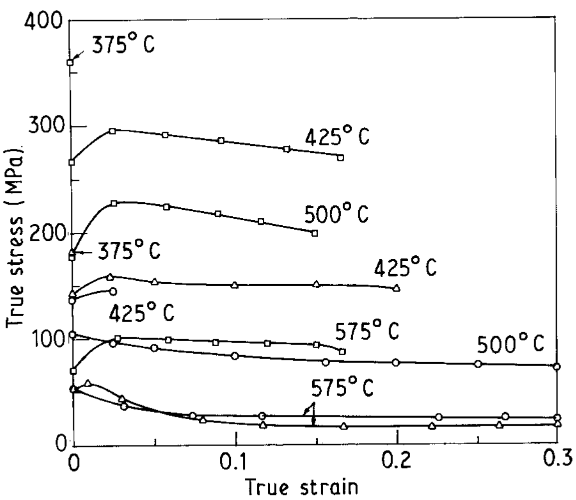

Figure 1 Stress–strain curves of different specimens ($\dot{\varepsilon} = 5 \times 10^{-5}\ s^{-1}$). (○) As-cast, (□) P/M, (△) single crystal.

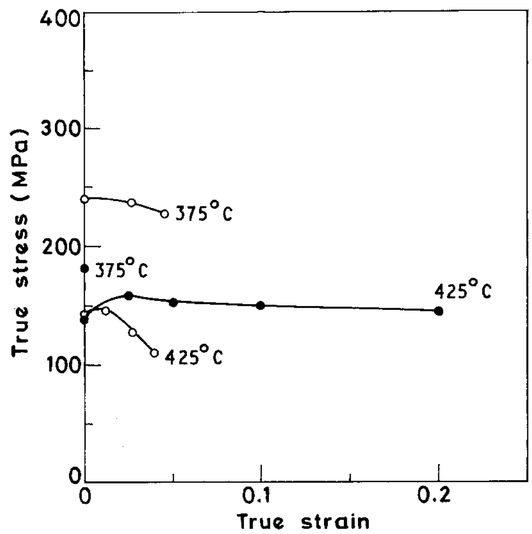

Figure 3 Stress–strain curves of single crystals (○) with and (●) without 3% prestrain at 475 °C. $\dot{\varepsilon} = 5 \times 10^{-5}\ s^{-1}$.

seen as a result of an increase in the initial dislocation density. For example, the as-cast specimen failed in a brittle manner at 400 °C, whereas the corresponding prestrained specimen exhibited extensive ductility at 375 °C. Thus the BDT temperature is lowered as a result of higher initial dislocation density.

The effect of temperature on flow stress for specimens of different grain size is shown in Fig. 5. It is noteworthy that the flow stress continuously decreases as the temperature up to nearly the melting point in $CuAl_2$. Another point of interest from the data of Fig. 5 is that the grain-size strengthening effect is retained over the whole temperature range up to its melting temperature. Moreover, the slope of the Hall–Petch plots is nearly independent of temperature (Fig. 6).

3.2. Flow mechanism

Assuming that a simple rate equation is valid for the

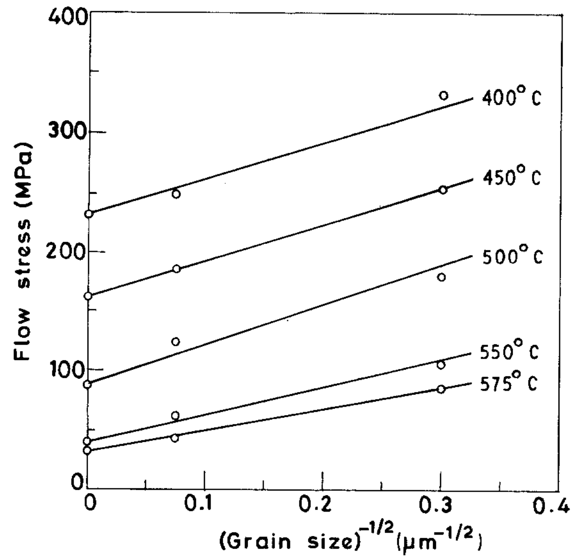

Figure 6 Plot of flow stress versus (grain size)$^{-1/2}$ at different temperatures.

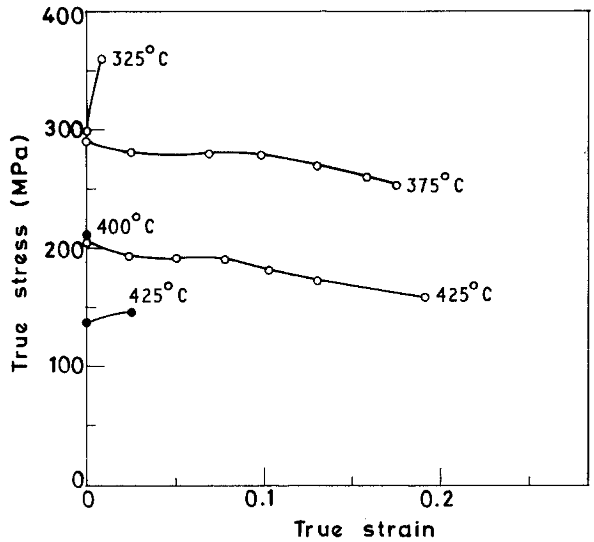

Figure 4 Stress–strain curves of as-cast specimens (○) with and (●) without 3% prestrain at 475 °C. $\dot{\varepsilon} = 5 \times 10^{-5}\,s^{-1}$.

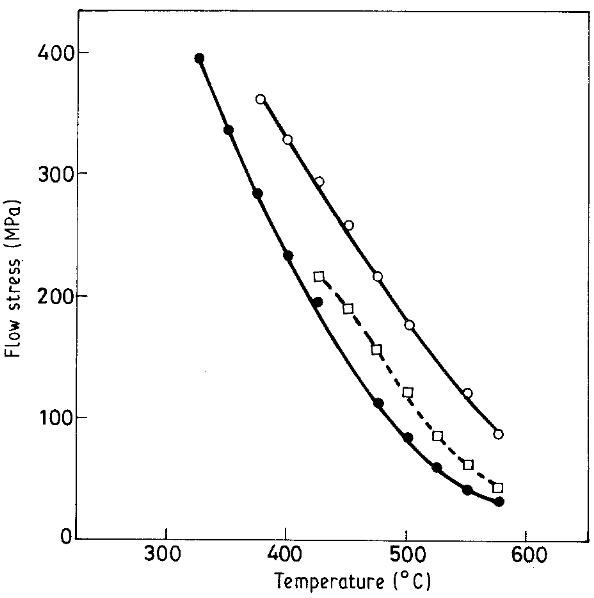

Figure 5 Plot of flow stress versus temperature for different specimens. (●) Single crystal, (□) Polycrystal (as-cast), (○) polycrystal (P/M).

thermally activated plastic flow, the activation parameters for plastic flow can be assessed experimentally and compared with the theoretical predictions for various flow mechanisms in order to identify the rate-controlling mechanism. The simple rate equation adopted for this purpose is

$$\dot{\varepsilon} = \dot{\varepsilon}_0 \exp - (\Delta H/kT) \qquad (1)$$

where $\dot{\varepsilon}$ is the plastic strain rate, $\dot{\varepsilon}_0$ the pre-exponential factor involving the number of activation sites for unit volume, frequency factor, etc., ΔH is the activation energy for plastic flow, and k and T have their usual meanings. Furthermore, the activation energy may be assumed to be linearly dependent on the effective stress, σ^*, as per the relation

$$\Delta H = \Delta H_0 - V\sigma^* \qquad (2)$$

where ΔH_0 is the total activation enthalpy of deformation including the work done by applied stress, V is the activation volume, and the effective stress, σ^*, is equal to the applied stress, σ, minus the athermal component of flow stress, σ_G. The activation volume, V, and activation energy, ΔH, are then given by

$$V = kT\left(\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma^*}\right)_T \qquad (3)$$

and

$$\Delta H = -kT^2\left(\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma^*}\right)_T\left(\frac{\partial \sigma^*}{\partial T}\right)_{\dot{\varepsilon}}$$

$$= -VT\left(\frac{\partial \sigma^*}{\partial T}\right)_{\dot{\varepsilon}} \qquad (4)$$

Making use of the strain-rate change test at a given temperature and temperature change test at a constant strain rate, the activation parameters assessed at different temperatures are shown in Fig. 7. While the activation volume is below 10 b^3 (where b is the atom size) in all cases, the activation energy is seen to vary linearly with temperature in the range of 375–525 °C. Furthermore, the activation volume at a given temperature is observed to be independent of plastic strain.

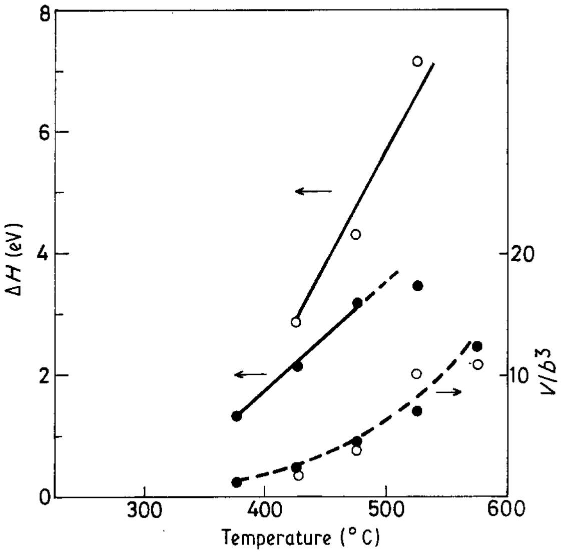

Figure 7 Activation energy and activation volume versus temperature plots for (●) single crystal and (○) polycrystal (P/M) specimens.

The above observations of strong temperature dependence of flow stress and small activation volume that is independent of plastic strain suggest that the Peierls mechanism is rate controlling in the plastic deformation of $CuAl_2$ up to 0.92 T_m, where T_m is the melting temperature. These observations are in agreement with the results of Dey and Tyson [9].

4. Conclusions

1. The brittle to ductile transition temperature of $CuAl_2$ is around 0.75 T_m and it is not significantly influenced by grain size.
2. The nature of grain boundaries depending on the cast or recrystallized conditions and the initial dislocation density varied by prior prestrain, affect the BDT temperature and the ductility.
3. The grain-size strengthening as per the Hall–Petch relation is valid up to a temperature close to the melting point of $CuAl_2$.
4. On the basis of thermal activation analysis of deformation, the rate-controlling process is identified as the Peierls mechanism up to 575 °C.

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Received 18 September 1991
and accepted 29 January 1992