flow stresses and the activation energy associated with yielding below 175°C. This cannot be done with the tensile test but it may be possible with compression tests, since it should be possible to reach much higher yield strength without brittle fracture.

The marked temperature and strain rate dependence of the yield strength of the body-centered-cubic metals at low temperatures is believed to arise from the anchoring of dislocations by interstitial solute atoms, such as carbon and nitrogen. There are at least three mechanisms by which these interstitial solute atoms can interact with dislocation. The first is by interaction with precipitated carbide or nitride particles. The second arises from the possibility of relief of hydrostatic stress by the solute atoms segregating into the dilated portion of edge dislocations. The third arises from the possibility of relief of shear stresses around both screw and edge dislocations by the solute atoms taking up preferred interstitial sites around the dislocation. The latter mechanism is believed to be the one chiefly responsible for the pronounced temperature and strain rate effects in the body-centered-cubic metals. The effects of strain rate and temperature on the yield strength and isostrain flow stresses could be correlated by the Zener-Hollomon parameter. The activation energy associated with yielding is 32,000 cal per mol and, within the limits of experimental error, is independent of strain.

The marked strain rate and temperature dependence of the yield strength of tungsten and of other body-centered-cubic metals is believed to be due to the interaction of interstitial solute atoms with dislocation arising from the relief of internal stresses by the solute atoms assuming preferred interstitial sites in the vicinity of the dislocations.

Discussion of this paper, if any, will appear in JOURNAL OF METALS, November 1956, and in AIME Metals Branch Transactions, Vol. 206, 1956.

References

Summary and Conclusions
The mechanical properties of annealed tungsten are extremely sensitive to strain rate in the temperature range 175° to 350°C, wherein the ductile to brittle transition occurs. The strain rate exponent, \( \tau_i \), in the equation \( \sigma_i \sim \dot{\varepsilon}^\tau \) is about 12 times as large as for steel.

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Interaction of Precipitation and Creep
In Mg-Al Alloys

Microstructural studies of a Mg-10.3 pct Al alloy showed that discontinuous precipitation during aging multiplies the grain boundary area available for easy deformation in elevated temperature creep. An increase of strain rate for a 6.2 pct Al alloy at 400°F, where precipitation and creep are concurrent, caused an increase in the volume percentage discontinuously precipitated at the completion of the process. The relatively poor elevated temperature creep resistance of heat treated alloys of the Mg-Al-Zn type was explained in these structural terms.

by C. S. Roberts

CREEP resistance of the heat treatable Mg-Al based alloys is poor compared to that of Mg-Ce based alloys. While the latter show a continuous precipitate, Mg-Al alloys prefer a discontinuous precipitate at grain boundaries. It has been concluded that the excellent creep resistance of Mg-Ce alloys was primarily due to the continuous precipitation which especially localized at the grain boundaries. The purpose of the present investigation was to obtain a parallel correlation, if such exists, between the poor creep resistance and the state of precipitation in Mg-Al alloys. Discontinuous precipitation produces about three times as many true grain boundaries as originally existed.

This microstructure would be expected to have a marked influence in creep, where quantity of grain boundary area is important.

Experimental Procedure
The two experimental alloys contained 6.2 and 10.3 pct Al. They were made from electrolytic magnesium by alloying under flux and casting in 3 in. diam permanent molds. The ingots were extruded into 1 1/4 x 1/4 in. flat stock under the following conditions: billet preheat 700°F (2 hr), container and die temperature 700°F, speed 3 fpm, and area reduction ratio 45:1. The creep specimens were machined from blanks which had been solution heat treated 24 hr at 770°F and aged 16 hr at 400°F. Attempts were made to control the grain size as closely as possible for extruded stock. The grain size after solution heat treatment was about two thousandths of an inch for the 6.2 and five thousandths for the 10.3 pct Al alloy.
The solution heat treatment temperature was selected to dissolve all the aluminum. The aging treatment was such that in the 10.3 pct Al alloy the volume of discontinuous precipitate had grown to the limit while in the 6.2 pct Al alloy more growth was possible during the creep test. The contrast between the microstructures is shown in Fig. 1. The 10.3 pct Al alloy is good for examining the influence of the final structure on the creep and the 6.2 pct Al alloy is suited for studying the influence of creep strain on concurrent precipitation.

The observations of microstructural changes were made on specimens which were electropolished and protected during testing by silicone oil. The procedure reported previously was used with one modification. Difficulties in electropolishing the aged alloy resulted from adherence of the precipitate as a fine gray powder on the specimen surface. This obstacle was circumvented by electropolishing in the solution heat treated condition followed by aging in a silicone oil bath. In order to avoid complicating the deformation structure with complete precipitate detail, these specimens were tested in the unetched condition.

Except for a few tests on the 10.3 pct Al alloy at constant load, the creep testing was performed under constant stress conditions with the aid of a simple noncollinear loading beam.

Influence of Discontinuous Precipitate on Creep

Microstructures of the 10.3 pct Al alloy after creep at 200° and 300° F showed evidence of the orientation relation developed by Smith. At these testing temperatures, slip is the principal deformation mechanism. Whenever slip lines were seen in all four regions involved at a grain boundary (the two original grains and the two recrystallized volumes), their direction was consistent with the predicted relation. Fig. 2 shows such evidence. The slip line evidence is sufficient to show that, in general, an orientation difference exists from region to region in an originally single grain.

Although this report is not intended to deal with individual strain-time relationships, it is well to include a brief description of the characteristic creep curves obtained from the 10.3 pct Al alloy in the range 200° to 600° F. Primary creep decreased in magnitude with increasing temperature. Secondary creep is of finite length at all temperatures. The tertiary stage begins at lower strain levels as temperature increases and always at less than 4 pct. Necking of the specimen does not occur until tertiary creep is well advanced.

The structure shown in Fig. 2 is characteristic of deformation at the low end of the temperature range studied. As in electrolytic magnesium, basal slip is profuse and finely spaced. The formation of low angle boundaries by kinking is present but there is no evidence of deformation twinning.

When the 10.3 pct Al alloy is tested at a high temperature and low stress, 600°F and 600 psi, deformation is highly localized at both the original and the precipitation-produced boundaries, Fig. 3. Inspection at X1000 failed to reveal slip lines. Sub-grain formation by kinking is very rare. Fig. 3 shows a smooth series of relief steps similar to those found in electrolytic magnesium. Apparently a cyclic process of boundary sliding and migration can operate in aged Mg-Al alloys at multitudes of new interfaces which result from the discontinuous precipitation process.

When approximately the same amount of strain occurred at 600°F and a higher stress, 1000 psi, sub-grain formation by kinking entered as an observable deformation process. A specimen strained to the same extent at 2000 psi shows slip in the microstructure, Fig. 4. The newly formed boundaries are primary deformation sites at 600°F.
The creep curves from the 6.2 pct Al alloy generally show the same form and temperature dependence as those from the 10.3 pct Al alloy. A relatively larger primary stage accompanied the occurrence of a small amount of mechanical twinning at lower temperatures and higher strain rates. A preferred orientation study showed that the maximum basal plane pole density exceeded 20° inclination to the extrusion surface normal.

The influence of the testing stress on discontinuous precipitation at 400°F when strain controlled is shown in Fig. 5. The data were obtained from specimens which were tested beyond the end of the race between the processes and generally well into the period of coalescence of overaged precipitate. As the stress is increased, the volume percentage which has discontinuously precipitated increases. This volume was quantitatively determined with the aid of a Hurlbut counter from electro-polished surfaces which were approximately halfway in depth between the original surface and the center of the specimen. The scatter of the data is reasonable in view of the human judgment required in such a quantitative metallographic study. The zero stress point was obtained from a specimen aged 6.2 pct Al alloy at 400°F.

Many examples of the acceleration of precipitation in general as a result of plastic deformation have been reported. Here is evidence that one type of precipitation is favored over another by the occurrence of plastic deformation. This is explained on the basis of build-up of elastic strain energy at the boundaries, favoring discontinuous precipitate formation. The unique appearance of concurrent discontinuous precipitate growth and deformation at the boundaries as they move is presented in Fig. 6.

Conclusions

1—The comparatively poor elevated temperature creep resistance of Mg-Al based alloys results from the ease of deformation at grain boundaries and the multiplication of these boundaries by discontinuous precipitation.

2—When the test temperature is such that precipitation is concurrent with creep, increasing stress or strain rate favors the discontinuous precipitate at the expense of the continuous type in Mg-Al alloys.

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References


Technical Note

Occurrence of CsCl-Type Ordered Structures in Certain Binary Systems Of Transition Elements

by Paul A. Beck, J. B. Darby, Jr., and O. P. Arora

LAVES and Wallbaum1 reported that the phases occurring at the compositions TiFe, TiRu, and TiOs, which are stable over a wide range of temperatures and are separated from neighboring phases by wide two phase fields, have a CsCl-type ordered structure. For TiRu and TiOs, this was recently confirmed by C. B. Jordan. Recent results of Greenfield and Beck2 show that solid solutions with ruthenium of the body-centered-cubic elements vanadium and tantalum (and presumably also of columbium) in group Va of the periodic table undergo CsCl-type ordering. Although detailed phase diagrams for these systems are not yet available, it would appear that these ordered structures are separated from the neighboring body-centered-cubic solid solutions by relatively narrow two phase fields.

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