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Many factors are known to affect the low temperature deformation characteristics of b.c.c. metals. One of the several controversial issues concerns the effects of interstitial impurities on the yield strength. Two different viewpoints have been suggested: (1) one group of investigators believes that interstitials, whether exclusively atomically dispersed or also partially segregated in precipitates, increase only the athermal component, τ_A, of flow stress and in no way alter the effective stress needed to assist thermal fluctuations in causing dislocations to surmount barriers (1-3); (2) a second group believes that increasing amounts of interstitial impurities increase the effective stress, τ^*, and thus alter the $\tau = \tau^* + \tau_A$ versus temperature T relationship for yielding (4-6). Furthermore, each group seems to hold alternate viewpoints on the nature of the low-temperature thermally activated mechanism of deformation: (1) investigators in the first group generally contend that the thermally activated mechanism arises from nucleation of pairs of kinks out of low Peierls valleys in general, or as modification thereof from a/2[111] screw dislocations having asymmetrically split cores (7-13); (2) the second group has generally felt that the rapid increase in the flow stress of b.c.c. metals with decreasing temperature arises principally from thermal activation of dislocations past high stress fields due to inter~~st~~itially positioned impurities (14-17).

Both viewpoints have undergone some modification as the result of more recent evidence that the Schmid law is not obeyed at low temperatures; the yield stress is asymmetric relative to orientation, and the tension and compression yield strengths differ from each other for the same orientation. It is the objective of this investigation to determine the effects of interstitial impurity content on the athermal stress, τ_A, and the effective stress, τ^*, of molybdenum single crystals oriented most favorably for $(\bar{1}01)[111]$ slip (Fig. 1).

Tensile specimens were made from 4 times electron beam zone refined Mo single crystal rods and the amount of interstitial impurity content was controlled by means of purification process (Tables I, II, III, IV).

Experimental results indicate that the athermal stress level, τ_A, is raised as the carbon content increases (Table V). While there is no certainty that all the carbon atoms are in solution, τ_A is approximately proportional (within the limits of experimental data) to the

TABLE I

Group A	C wt.%	O wt.%	N wt.%	Treatment and Description
1	.017	.012	.009	Four zone refining passes. Metallographic observation showed no precipitation.
2	.006	.007	.002	
3	.015	.005	.003	
4	.012	.007	<.001	
5	.023	.011	.002	
6	.022	.012	.002	
7	.020	.014	.002	
8	.006	.014	.006	

TABLE II

Group B	C wt.%	O wt.%	N wt.%	Treatment and Description
1	.013	.002	.002	Four zone refining passes. 4% H_2-96% He annealing at 2000°C for 24 hours. Furnace cooled, temperature fell to below 700°C in less than 2 minutes. Metallurgraphic observation showed no precipitation in the interior of the specimens.
2	.008	.002	.001	
3	.012	.002	.003	
4	.006	.002	.005	
5	.015	.002	.006	
6	.007	.002	.006	
7	.013	.002	.014	
8	.013	.002	.008	
9	.006	.002	.006	
10	.009	.002	.002	
11	.006	.002	.002	
12	.008	.002	.001	

TABLE III

Group C	C wt.%	O wt.%	N wt.%	Treatment and Description
1	.009	.003	<.001	Four zone refining passes. H_2 annealing at 24 hours at 2000°C. Five hours of cooling time to room temperature. No surface ppt. Metallographic observation and electron microprobe detected carbide ppt in the interior of the specimens.
2	.010	<.002	<.001	
3	.010	.003	<.001	
4	.008	<.002	<.001	
5	.011	<.002	<.001	
6	.010	<.002	<.001	
7	.005	<.002	<.001	
8	.006	.002	.001	
9	.005	.002	.002	

square root of carbon content $\sqrt{C}$ or carbon content C. The effective stress, τ^*, versus temperature relationship for all four different impurity content remains the same (Fig. 2); furthermore,

TABLE IV

Group D	C wt.%	O wt.%	N wt.%	Treatment and Description
1	<.002	.002	.003	Four zone refining passes. 11 days of ZrH_2 purification at 1010°C. Five hours cooling time to room temperature. No surface ppt.
2	<.002	.002	.005	
3	<.002	.002	.001	
4	<.002	.003	.003	
5	.002	.002	.002	
6	<.002	.003	.001	
7	<.002	.002	.001	

Chemical analysis was performed by the Coors Spectrographic Laboratory.

TABLE V

Group	0.2% Offset Athermal Stress at 550°K	Average Carbon Conc. Wt.%
A	2.65×10^8 dynes/cm^2	0.015
B	2.26×10^8 dynes/cm^2	0.0097
C	1.62×10^8 dynes/cm^2	0.0082
D	1.48×10^8 dynes/cm^2	<0.002

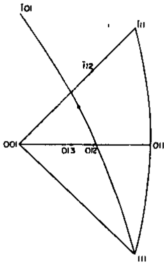

FIG. 1
Orientation of the tensile axis of the crystal.

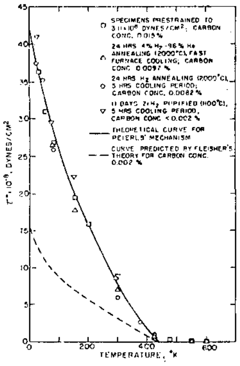

FIG. 2
Thermally activated component of flow stress vs. temperature.

this relationship agrees closely to that ascribed to the Peierls mechanism (8) (Figs. 2 and 3). Fleischer's theory (14) of interactions between dislocations and tetragonal distortion strain centers would predict different temperature dependence of flow stress for specimens containing different interstitial impurities. If the temperature dependence of flow stress of group 1 specimens (0.015% C) was ascribed by the solid line in Fig. 2, then according to Fleischer's theory, the temperature dependence of group 4 specimens (<0.002% C) should have followed the dotted line. It is shown in Figs. 2 and 3 that all four groups of specimens have the same effective stress versus temperature relationship. Activation volume experiments indicate that for four different groups of specimens the apparent activation volumes, Va, are small (5-60 b^3), relatively independent of stress and of impurity content. Furthermore, the order of magnitude and their stress dependence agree well with those predicted by the Peierls theory (8) (Figs. 4 and 5).

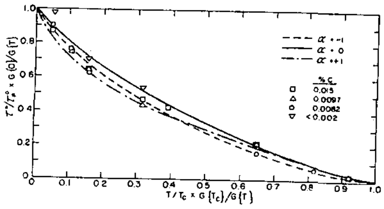

FIG. 3
Thermally activated component of flow stress vs. temperature in dimensionless units.

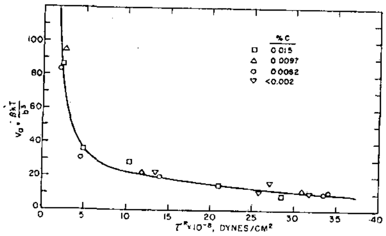

FIG. 4
Thermally activated component of flow stress vs. apparent activation volume.

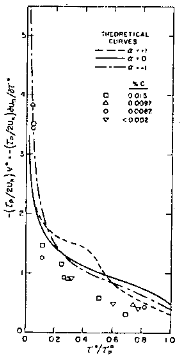

FIG. 5
Thermally activated component of flow stress vs. activation volume in dimensionless units.

In view of all experimental evidence, it is concluded that the interstitial impurity content only affects the athermal stress, τ_A, and the Peierls theory remains as the most attractive model to account for the strong temperature dependence of flow stress of b.c.c. metals. A modified model for thermally-activated deformation in b.c.c. metals (18) has been formulated recently for the rationalization of the asymmetric plastic behavior of single crystals of b.c.c. metals. Experiments are now in progress to investigate the stress-strain rate-temperature dependence of yield strength of b.c.c. metals as a function of orientation.

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