THE EFFECT OF ORIENTATION, TEMPERATURE AND GAMMA PRIME SIZE ON THE YIELD STRENGTH OF A SINGLE CRYSTAL NICKEL BASE SUPERALLOY

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Summary

The results of a detailed study of the effects of orientation, temperature, and γ' size on the yield strength of a single crystal nickel base superalloy, PWA 1480, are presented. At 593°C (1100°F) orientations along the <001> to <011> boundry of the standard stereographic triangle deform by octahedral slip, but do not obey Schmid's law. Moreover, the deviation from Schmid's law is asymmetrical in tension and compression. The <001> orientation is stronger than the <011> orientation in tension, but the reverse is true in compression. Orientations near <111> were found to deform by homogeneous cube slip. The yield strength exhibits a plateau between room temperature and 760°C (1400°F) for specimens with fine γ' size. Coarsening of the γ' leads to a significant drop in the yield strength at room temperature for the <001> orientation. Near the peak temperature of 760°C (1400°F) the yield strength is insensitive to γ' size. Thus, the behavior of material with a coarse γ' size approaches that of an $L1_2$ compound, showing an anomalous increase in strength with increasing temperature. For the <111> orientation, however, coarsening of γ' causes a uniform decrease in strength at all temperatures. Many of these observations are consistent with the behavior of $L1_2$ compounds and can be explained by the activated cube cross-slip mechanism proposed for $L1_2$ compounds. The plateau in yield strength for specimens with a fine γ' size is attributable to simple geometrical hardening.

Introduction

In recent years, many basic studies have been carried out to better understand the deformation behavior of single crystals with the $L1_2$ structure (1-3). The motivation behind these studies has been the high temperature application of precipitation hardened nickel-base superalloys which are hardened with $L1_2$ compounds based on Ni_3Al. However, few attempts have been made to relate the published work on $L1_2$ compounds to the understanding of the temperature dependence of the deformation behavior of nickel-base superalloys (4,5). With the application of single crystal superalloy technology to the gas turbine engine (6), there is a real motivation to bridge this gap. This study represents an effort in that direction.

The most useful and interesting aspect of the deformation behavior of single crystal $L1_2$ compounds is the anomalous increase in strength with increasing temperature for the cube orientation. This unusual behavior has been attributed to a thermally activated cube cross-slip process inhibiting dislocation glide. The cross-slip process also leads to a tension/compression asymmetry in flow stress (3). It will be shown that many aspects of the deformation behavior of single crystal superalloys can be understood in terms of the cube cross-slip mechanism.

Experimental Procedure

Standard tensile specimens with 10 mm gauge lengths and 5mm x 5mm x 15mm compression specimens were machined from single crystal PWA 1480 (5 Co, 10 Cr, 4 W, 12 Ta, 5 Al, 1.5 Ti, balance Ni). As-cast bars were solution heat treated for four hours at 1288°C (2350°F) and air cooled. This was followed by a four hour age at 1079°C (1975°F) and a 32 hour age at 871°C (1600°F). To obtain specimens with various γ' sizes, the solution heat treatment cycle was modified both with respect to temperature and cooling rate. Some of the compression specimens were electropolished prior to testing in order to make slip line observations. All tensile and compression tests were carried out at a strain rate of 0.5 percent/minute.

Results

Orientation and Temperature Dependence of the Yield Strength

The yield strength of PWA 1480 as a function of temperature for the three major orientations, <001>, <011> and <111>, in unidirectional tension and compression, is presented in Figure 1. The results provide an overview of the temperature and orientation dependence of the yield strength of single crystal PWA 1480. All orientations show a plateau in yield strength between room temperature and 760°C (1400°F) with a slight dip at around 400°C (750°F). A significant drop in strength occurs beyond 760°C (1400°F). At temperatures above 400°C (750°F) the <111> is the weakest orientation at all temperatures in both tension and compression. While the <001> orientation is stronger than the <011> orientation in tension, the reverse is true in compression. This tension-compression asymmetry is most pronounced between room temperature and 760°C (1400°F).

To better understand the nature of the orientation dependence of the yield strength, specimens in various orientations were tested both in tension and compression at 593°C (1100°F). These results are presented in Figure 2(a) and (b). The relative tensile versus compressive strength as a function of orientation is schematically represented in Figure 2(c). Orientations near <001> are stronger than orientations near <011> in tension, but the reverse is true in compression. This can be better seen in a plot of yield strength versus orientation where orientation is measured as the angle from the <001> direction going toward the <011> direction, as depicted in Figure 3. The plot indicates the failure of Schmid's Law. Based on the highest resolved shear stress criterion, both the <001> and <011> orientations would be expected to deform by octahedral slip. As both orientations have the same Schmid factor they should have equivalent yield strengths. Figure 3 shows that this is not the case and that the deviation occurs in both tension and compression in opposite senses.

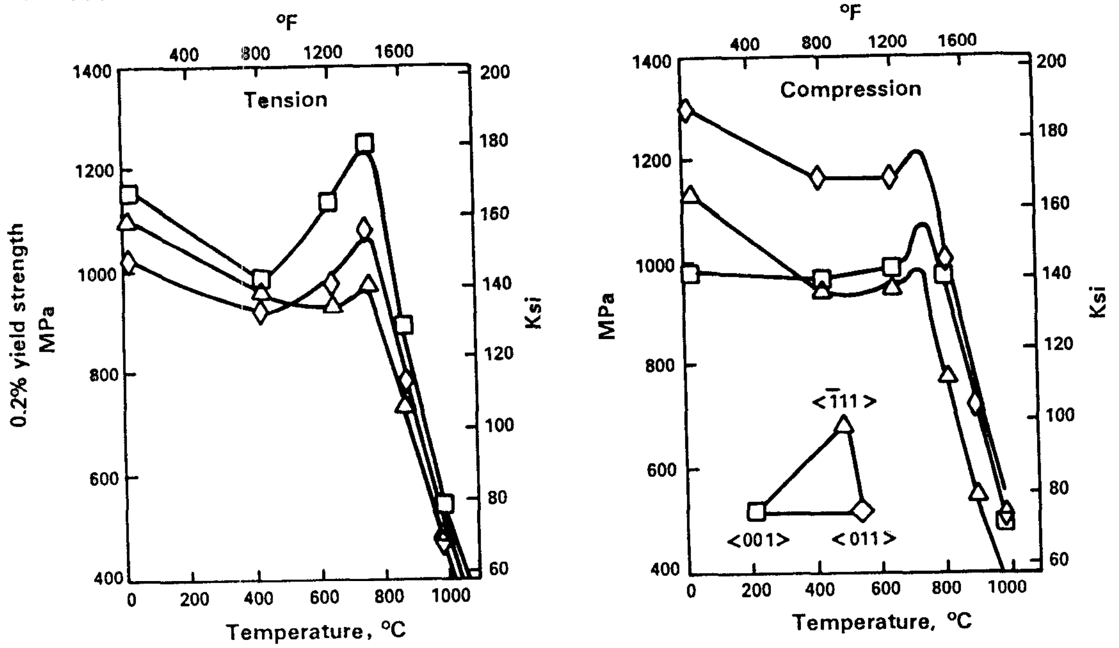


Figure 1 -Yield strength of PWA 1480 as a function of temperature for the <001>, <011> and <111> orientations in tension and compression.

To study slip behavior, compression specimens in various orientations, with electropolished faces, were strained 0.5 percent at 593°C (1100°F). The <001> orientation was observed to deform by inhomogenously distributed octahedral slip while the <111> orientation deformed by homogenously dispersed cube slip. Two other orientations, <011> and <124>, showed slip behavior akin to that of the <001> orientation with patches of slip traces indicative of cross-slip activity.

Effect of γ' Size On Yield Strength

Specimens with γ' sizes varying from 0.3 microns to 3.0 microns were tested in compression between room temperature and 870°C (1600°F). The results for the <001> and <111> orientations are plotted in Figure 4(a) and 4(b), respectively. At about 760°C (1400°F) the yield strength in the <001> direction is insensitive to the γ' size compared to the variation in strength at room temperature. A <001> oriented PWA 1480 single crystal with coarse γ' behaves like single phase $Ni_3(Al,X)$, displaying an anomalous increase in yield strength with increasing temperature. However, for the <111> direction, consistent with the behavior of $Ni_3(Al,X)$, no anomaly is observed for PWA 1480 with a coarse γ' size. For the <111> orientation the increase in yield strength with decreasing γ' size is more or less uniform at all temperatures.

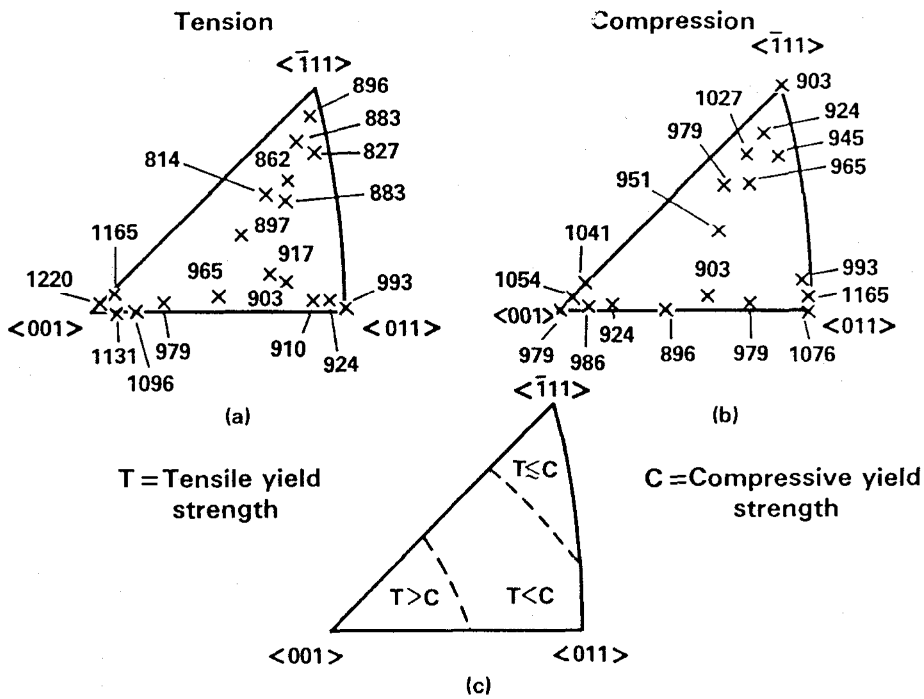


Figure 2 - Orientation dependence of the yield strength in MPa at 593°C (1100°F) for PWA 1480: (a) in tension, (b) in compression, and (c) schematic diagram showing tension/compression asymmetry.

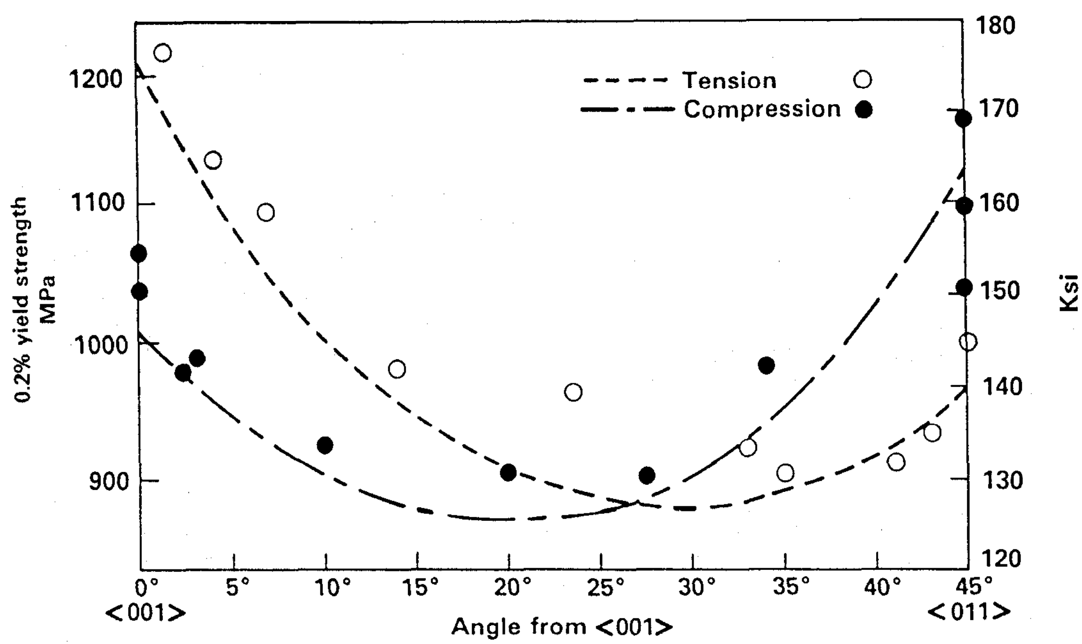
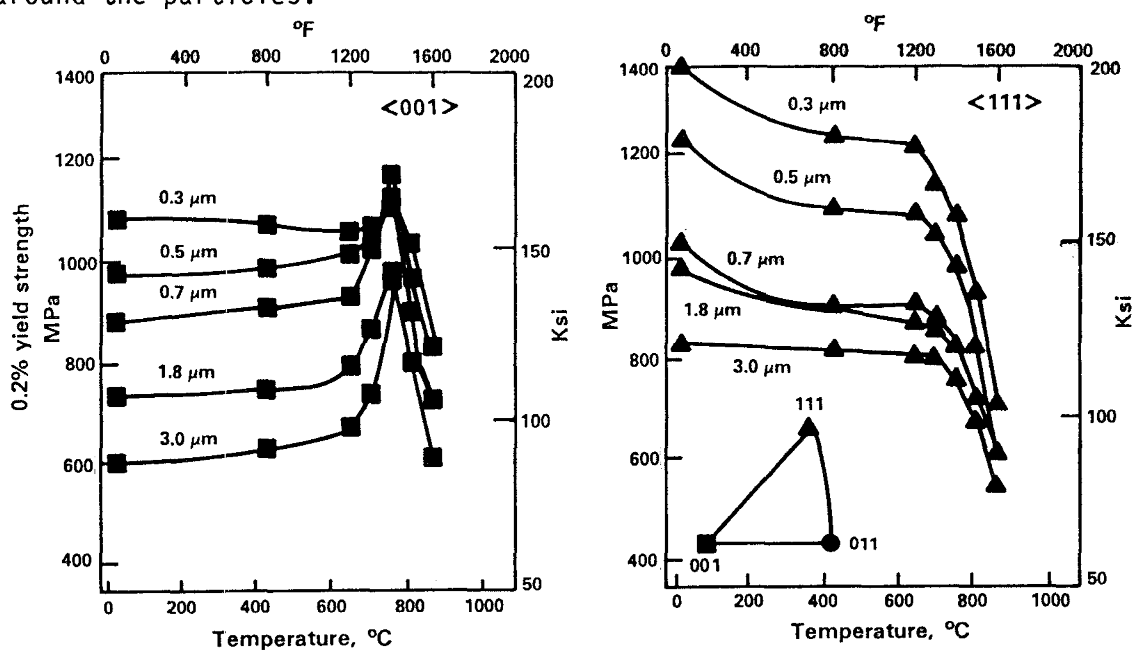


Figure 3 - Yield strength of PWA 1480 at 593°C (1100°F) versus orientation along the <001> - <011> boundary of the standard stereographic triangle.

Slip trace analysis was carried out for specimens strained at 649°C (1200°F) with the finest and the coarsest γ'. Consistent with the observations for $L1_2$ compounds, the slip was homogeneous for PWA 1480 with coarse γ'. However, for PWA 1480 with a fine γ' size the slip was highly inhomogeneous. Cube slip, which is normally difficult to observe because of its homogeneous nature, was readily observable for the <111> oriented specimen with the finest γ' size.

Discussion

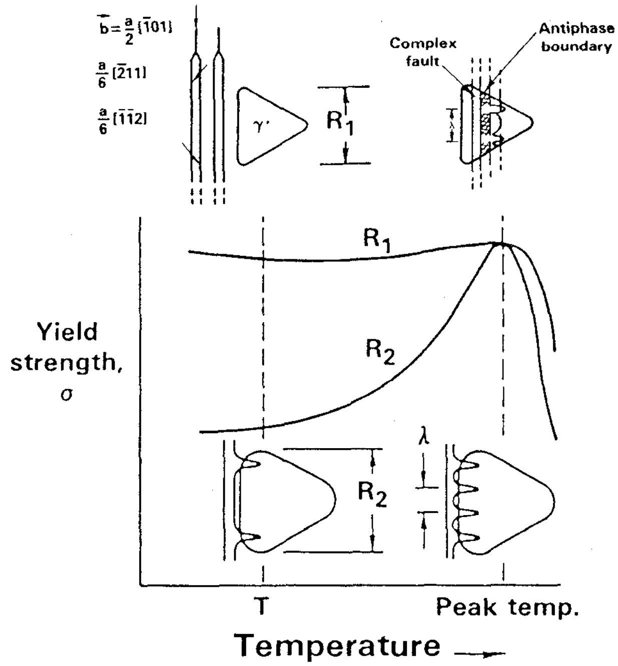
From the results presented herein, three interesting aspects of the deformation behavior of single crystal PWA 1480 emerge: 1) Schmid's law is not obeyed, 2) an asymmetry exists in the tension/compression yield strength and 3) the yield strength at intermediate temperatures near 760°C (1400°F) is insensitive to γ' size. It can be shown that all these characteristics of the yield strength behavior of single crystal PWA 1480 can be rationalized based on the activated cube cross-slip model developed for $L1_2$ compounds (1,2), provided it is assumed that γ' shearing is the primary strengthening mechanism. In the case of superalloys with high volume fractions of γ' (>50 volume percent), this is a justifiable assumption. With the mean free edge-to-edge distance in the matrix between the precipitates being smaller than the average precipitate size itself, dislocation shearing of the particles is favored over dislocation looping around the particles.



(a) (b)

Figure 4 - Effect of γ' size on the yield strength of PWA 1480 in: (a) <001> orientation, and (b) <111> orientation.

Obviously, the γ' shearing mechanism requires a pair of dislocations so that a large antiphase boundary (APB) is not created. However, as soon as the dislocation pair enters an ordered γ' precipitate, cube slip becomes a viable slip mode and, if sufficient thermal activation is available, small segments of the leading dislocation cross-slip to form sessile obstacles (1). As shown schematically in Figure 5, it is the mean free distance, λ, between the cross-slip events that becomes the strength controlling parameter. This rather simple model of the γ' shearing mechanism can be used to qualitatively explain the temperature dependence of the yield strength for single crystal PWA 1480 in the <001> orientation with varying γ' size, as shown in Figure 5.



$$\sigma \propto \left(\frac{1}{R}+\frac{1}{\lambda}\right)$$

Figure 5 - Yield strength as a function of temperature for a single crystal superalloy with fine and coarse γ' deforming by octahedral slip.

If we assume that at room temperature the mean free distance, λ, is smaller than the γ' precipitate size for PWA 1480 with a coarse γ' size and hence the strength controlling parameter (lower curve in Figure 5), with increasing temperature, enhanced thermal activation leads to a greater propensity for cross-slip and a smaller λ. This results in a higher γ' shearing stress and an anomalous increase in yield strength with increasing temperature. For PWA 1480 with a fine γ' size the mean free distance, λ, between the cube cross-slip events is smaller than the γ' size only at high temperatures (upper curve in Figure 5). At lower temperatures λ is no longer the strength controlling factor since at any temperature for any γ' size, it is the smaller of the two parameters - the γ' size, R, or the mean free distance, λ, between the cross-slip events that controls the strength. Thus, for PWA 1480 with fine γ' a plateau in strength results between room temperature and the peak temperature since the γ' size controls the strength. At the peak temperature, however, irrespective of the γ' size, the mean free distance between the cross-slip events becomes very small and controls the strength. Consequently, the yield strength is insensitive to variations in γ' size.

The above discussion was intentionally limited to deformation of the <001> orientation, corresponding to results presented in Figure 4(a), since it was tacitly assumed that deformation took place by octahedral slip. Obviously, for the <111> orientation, deforming by cube slip, the process of thermally activated cross-slip is irrelevant. In this case, with very high resolved shear stresses acting on the cube plane, the entire dislocation segment cross-slips and shears the γ'. Thus, the geometrical limitation of γ' size is the controlling parameter at all temperatures. This is reflected in Figure 4(b) where strength at all temperatures is uniformly raised upon refining the γ' size.

So far we have considered only thermal activation as the driving force for the formation of sessile cross-slipped segments on the cube plane. In addition to thermal activation, interaction of the dislocation core structure with the applied stress tensor is also an important factor controlling the propensity for cross-slip (1-3). We consider this by recognizing that the cross-slip process is aided by a higher resolved shear stress for the cross-cube plane. In addition since the dislocation partials bounding a stacking fault must come together before cross-slip can occur, stress aiding or retarding that process would also affect cross-slip.

Contours of the Schmid factor for primary octahedral slip (111) $[\bar{1}01]$ are presented in Figure 6(a) where the implication is that the higher the Schmid factor the lower the yield strength - provided that the critical resolved shear stress (CRSS) for octahedral slip is not a function of orientation. Figure 6 (b) and (c) show contours for the Schmid factor ratios for cube cross-slip, N, and the constriction stress, Q. The parameter N is a measure of the shear stress for cube cross-slip. The region in Figure 6(b) with N>1 represents orientations which will deform by cube slip if the CRSS for both octahedral and cube slip were identical. This is the region near the <111> orientation. For orientations where N<1, higher values of N indicate a greater probability for cross-slip. Thus in going from the <001> to the <011> orientation, as N increases, the tendency for cross-slip is expected to increase as does the CRSS for octahedral slip. This is a partial rationalization for the <011> orientation having a higher yield strength than the <001> orientation in compression, as shown in Figure 3. Ideally, if the CRSS were not a function of orientation, both <001> and <011> orientations should have equal yield strengths based on equivalent Schmid factors.

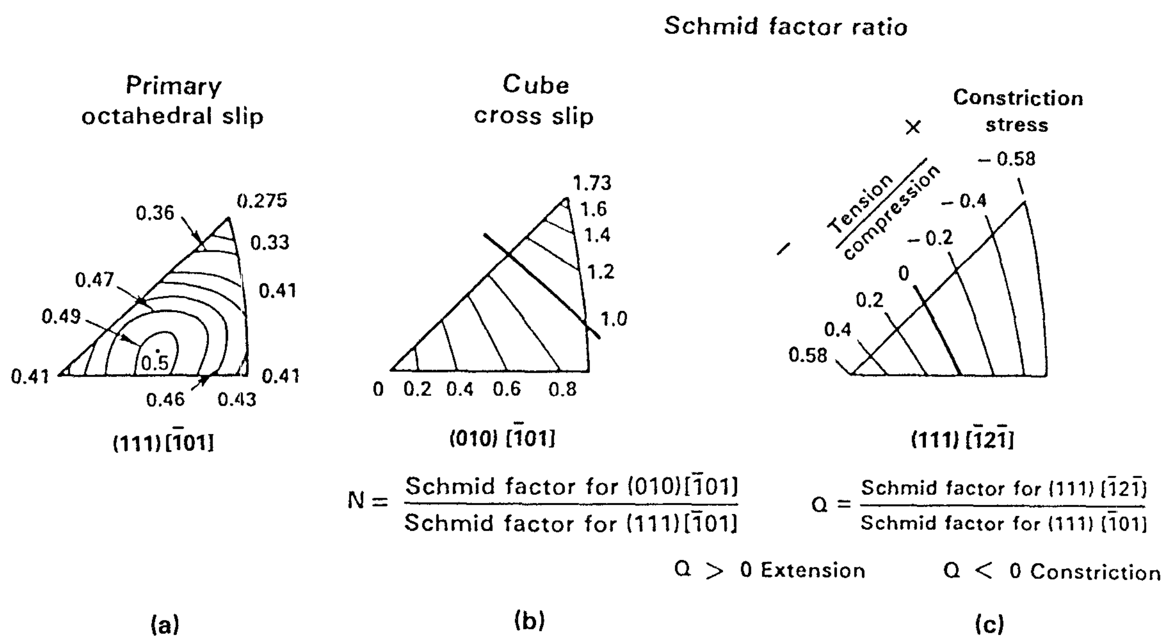


Figure 6 - (a) Contours of Schmid factor for primary octahedral slip (b) contours of Schmid factor ratio for cube cross-slip, and (c) contours of Schmid factor ratio for constriction stress.

It can be seen from Figure 3 that the above argument fails to explain the yield strength behavior in tension, where the <001> orientation is stronger than the <011> orientation. To understand this reversal in behavior with the change in sign of the stress we must consider the resolved constriction stress, which for the <001> orientation is shown in Figure 7. The direction of the glide force per unit length, Fg/L, will reverse upon reversing the applied stress, σ. While this has no physical significance in a macro sense, since it only alters the direction of glide for a particular dislocation, reversing the direction of the glide forces acting on the dislocation partials leads to a distinctly different physical situation. The resulting force tends to constrict the partials under an applied tensile stress and extend them under a compressive stress. Since constriction of the partials aids the cross-slip process, the <001> orientation appears stronger in tension than in compression where the extended partials retard cross-slip activity. Thus for the parameter Q not only the magnitude, but also the sign is critical, which reverses upon reversing the sign of the applied stress. This fact is reflected in Figure 6(c). Physically, Q = 0 means no effect on cross-slip due to the constriction stress and hence no asymmetry in yield strength between tension and compression. The similarity between Figure 2(c) and Figure 6(c) is obvious suggesting that the crossover in the yield strength asymmetry shown in Figure 2(c) probably occurs around Q = 0 as shown in Figure 6(c).

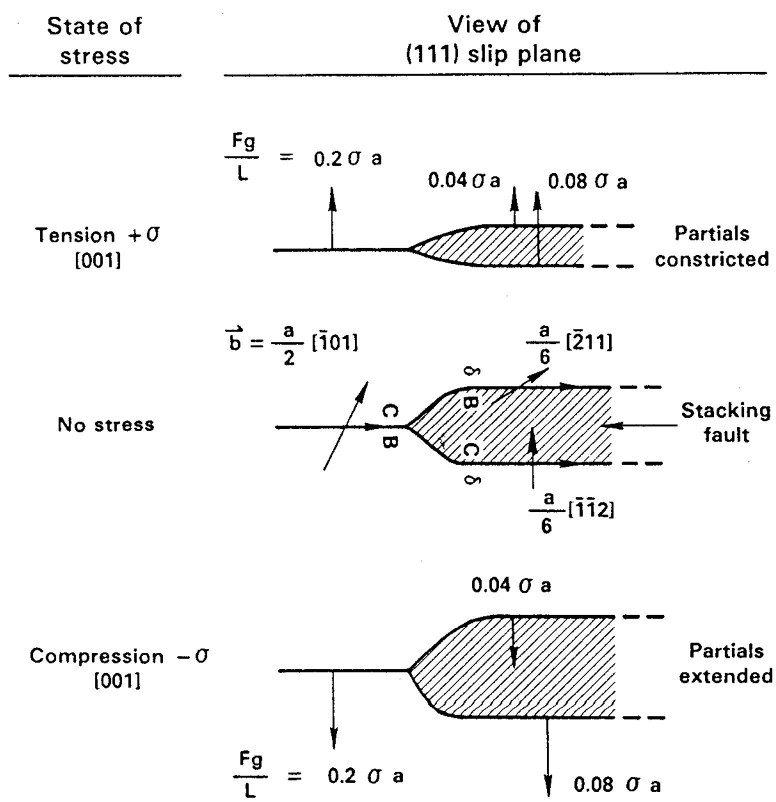


Figure 7 - Effect of constriction stress on a <001> oriented crystal under tension and compression. (a is the lattice parameter)

The preceeding discussion has been limited to temperatures below about 760°C (1400°F). The drop in strength at higher temperatures has been attributed to various factors, which include, the possibility of a reverse cross-slip reaction (9), introduction of cube slip and reduction of the APB energy. However, it is unlikely the APB energy has any effect on <111> orientations deforming by cube slip or that cube slip operates for the <001> orientation. In addition the strain rate sensitivity of the yield strength in the high temperature region must also be accounted for (10).

The preceeding discussion may be summarized quantitatively with the following equation:

$$\sigma_y.m = A\,\frac{\gamma(T)}{b} + B\,\frac{Gb}{R} + C\,\exp\frac{(-H_o+V_1N\sigma_y m-V_2Q\sigma_y m)}{kT}$$

The CRSS, $\sigma_y.m$, which is the yield strength times the Schmid factor, m, has three components. The creation of APB upon shearing the γ' particles is reflected by the term $A\gamma(T)/b$, where $\gamma(T)$ is the APB energy as a function of temperature (T), b is the Burgers vector, and A is a constant. The second term, which is inversely proportional to the γ' size, R, represents the geometrical hardening term, G is the shear modulus of $\gamma'$and B a constant. The last term is a measure of the strengthening contribution due to the cube cross-slip process inhibiting dislocation shearing of the γ' precipitates. In this term, H_o is the thermal activation energy and V_1 and V_2 are activation volumes for cube cross-slip and constriction stress interactions, respectively. N and Q are Schmid factor ratios as previously defined, k is Boltzmann's constant and C is a constant.

From the equation, it is clear that at low temperatures, for superalloys with coarse γ', the yield strength is primarily controlled by the APB energy. For superalloys with fine γ' the yield strength is primarily a result of the APB energy and the geometrical hardening term. At the peak temperature, however, the yield strength is predominantly controlled by APB energy and thermally activated cross-slip.

Conclusions

1. Superalloy single crystals such as PWA 1480 with a high volume fraction of γ' deform by octahedral slip near the <001> orientation and by cube-slip near the <111> orientation.

2. Shearing of the γ' precipitate particles is the principal strengthening mechanism.

3. Strength at the peak temperature is insensitive to γ' size for <001> oriented superalloys due to the strong contribution of thermally activated cube cross-slip.

4. Schmid's law is not obeyed. The critical resolved shear stress is altered by cube cross-slip and the constriction stress acting on the partial dislocations.

5. Orientations near the <001> - <011> boundary of the standard stereographic triangle show a tension/compression asymmetry consistent with the contribution of the constriction stress to the cube cross slip process.

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