Effect of Annealing and Etching on Machining Damage in Structural Beryllium

JAMES E. HANAFEE

Machining damage can seriously degrade the mechanical properties of beryllium. In this study substantial machining damage was introduced in two different structural grades of beryllium. Two methods of relieving the machining damage were studied: annealing at various temperatures and etching away various amounts of surface material. For the depth of damage introduced in the study (approximately 100 μm (0.004 in.)), removing the machining damage in the first 25 to 50 μm (0.001 to 0.002 in.) by either etching or annealing restored the mechanical properties of the beryllium. Microhardness was useful for ascertaining the extent of machining damage.

INTRODUCTION

There have been substantial advances in beryllium metallurgy in the 1970's: isotropic ultimate strengths over 550 MPa (80,000 psi) and yield strengths over 350 MPa (50,000 psi) can now be achieved while not only maintaining ductility but also actually raising elongation at fracture to over 3 pct for all directions. These properties can be obtained reproducibly and on a production scale for power-origin beryllium.

However, beryllium is quite susceptible to machining damage.[1-8] The damage is in the form of microstructural changes that take place during the machining operation, and the mechanical properties of the metal can be seriously degraded. For example, the ultimate strength can be decreased as much as 20 to 30 pct and elongations can be dropped from over 3 pct to less than 1 pct. Thus, it may be quite difficult to take advantage of the major improvements in beryllium of the past few years unless machining damage is avoided or removed.

Traditionally there have been three approaches for alleviating the machining-damage problem. The first approach is to tightly control machining procedures, particularly the depth of cut. This can be successful and, indeed, it has been shown that a Materials Advisory Board (MAB) recommended procedure[19] that entails shallow cuts can result in no machining damage and no degradation of mechanical properties.[15] However, King has shown that a 25-μm-(0.001-in.-) deep cut caused about 500 μm (0.020 in.) of machining damage.[13] Thus, controlling the depth of cut does not ensure freedom from machining damage, and because there are no practical nondestructive tests for detecting machining damage, this approach would not ensure the reliability of machined beryllium parts.

The second approach for alleviating machining damage is to anneal the machined part. This has the disadvantage of possibly introducing distortion, and some steps in the manufacturing of specific beryllium parts may preclude such an approach. Another possible disadvantage of annealing is that most beryllium grades are in fact microalloys, and it is quite conceivable that annealing will change mechanical properties beyond just relieving degradation caused by machining damage.

The third approach is to chemically remove the

JAMES E. HANAFEE, Lawrence Livermore Laboratory, University of California, P.O. Box 808, Livermore, CA 94550.

ISSN 0162-9700/80/0706-0041$00.75/0

damaged surface layer from the beryllium part. This approach is effective but has the disadvantages of possibly violating dimensional tolerances and creating a waste-disposal problem. Also, as for the annealing approach, some steps in manufacturing may militate against such an approach.

A deficiency of some machining-damage studies has been that machining damage was usually not introduced to the degree necessary to affect mechanical properties consistently and to a sufficiently large degree. Further, studies often have not quantitatively compared the relative merits of chemical removal and annealing for different grades of beryllium. In this study machining damage was purposely introduced in tensile specimens from two very different, useful grades of structural beryllium. In this study, annealing at various temperatures and etching to various depths were employed to relieve the damage. The resulting effects on tensile properties, microstructure, and microhardness were investigated.

EXPERIMENTAL PROCEDURES

Two structural grades of powder-origin beryllium were selected for the study. One grade, Brush-Wellman S200E, is representative of medium-strength and medium-purity material that has been used for many years (Table I). The second grade, referred to as "1319A", was purchased according to a Lawrence Livermore Laboratory specification,[20] which requires higher mechanical properties and lower impurity levels. The material that met the Lawrence Livermore Laboratory (LLL) specification was fabricated using advances in beryllium technology that were made during the 1970's, and in several respects it is similar to the commercially available S65, manufactured by Brush-Wellman and to the formerly commercially available CHIP-HIP 1, manufactured by Kawecki-Berylco Industries (KBI).*

* Reference to a company or product name does not imply approval or recommendation of the product by the University of California or the U.S. Department of Energy to the exclusion of others that may be suitable.

Several machining variables were explored in order to introduce a consistent and reasonable magnitude of machining damage into the beryllium tensile specimens —specifically, the cutting speed, depth of cut, feed rate, and tool radius. The final choices were: a cutting speed of 84 rad/s (800 rpm), a feed rate of 0.42 mm/s (1 in./min), a depth of cut (on the last two passes) of 0.65 mm (0.025 in.), and a tool radius of 0.81 mm (0.032 in.). These were chosen on the basis of the resulting microstructural change and degradation of mechanical properties.

In the annealing experiments, the time for the specimens to reach the annealing temperature ranged from 15 min for the low-temperature (400 °C (725 °F)) to about 2 h for the high temperature (1000 °C (1832 °F)).

Table I. Certified Chemical Analysis and Mechanical and Physical Properties of the Two Beryllium Grades Used in this Study

Chemical analysis, wt pct

Grade*	Be (assay)	BeO	Fe	Al	C	Mg	Si	Ni	Cr	Mn	Cu	Ti	S	U
S200E†	98.8	1.3	0.10	0.05	0.11	0.01	0.02	—	—	—	—	—	—	—
1319A	98.9	1.19	0.095	0.006	0.050	0.008	0.022	0.016	0.005	0.004	0.007	0.003	<0.015	0.006

Mechanical properties

	Longitudinal			Transverse		
Grade	Ultimate strength, MPa (ksi)	0.2 pct-yield strength, MPa (ksi)	Elongation, Pct	Ultimate strength, MPa (ksi)	0.2 pct-yield strength, MPa (ksi)	Elongation, Pct
S200E	319 (42.3)	224 (32.5)	1.6	366 (53.1)	227 (32.9)	3.7
1319A‡	353 (51.2)	259 (37.6)	3.2	386 (56.0)	273 (39.6)	6.2

Physical properties

Grade	Density, g/cm^3	Grain size, μm
S200E	1.853	18.2
1319A	1.854	11.3

*Specification S200E; Brush-Wellman, producer; BW pressing 0819. Specification MEL76-001319A: Kawecki-Berylco Industries, producer; KBI unit 517A-1.

†Less than 0.04 wt pct metallic impurities other than specific analysis.

‡Mechanical-property values are averages of tensile specimens from top and bottom of pressing.

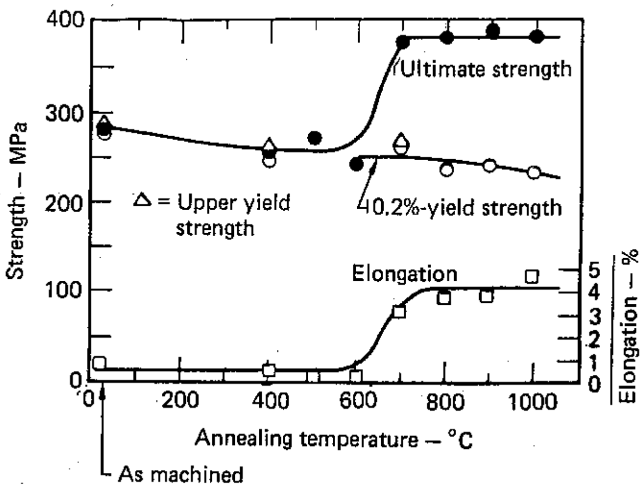


Fig. 1—Effect of 1-h anneal at various temperatures on the mechanical properties of machining-damaged 1319A beryllium.

Cooling was accomplished in vacuum by removing the furnace, and the tensile specimens cooled to below 200 °C (392 °F) in less than 20 min from 1000 °C (1832 °F) and in less than 5 min from 400 °C (752 °F).

In the etching experiments, the specimens were etched at 55 ± 5 °C (132 ± 9 °F) in a bath of 225 ml H_3PO_4, 45 ml H_2O, 9 ml H_2SO_4 and 25 g CrO_3. The etch was followed by rinsing with hot water, cold water, and finally alcohol. Mechanical testing was carried out according to the MAB recommendations.[19] The specimens were 6 mm (0.25 in.) in diam and had a 25.4-mm (1-in.) gage length.

The tensile specimens were also used for the microscopy and microhardness studies. The relatively light load (50 g) used for the microhardness studies introduced some variation in individual readings for some specimens (up to about ± 15 VPH) but also permitted a more accurate assessment of the rapidly changing hardness in the narrow machining-damaged region.

RESULTS AND DISCUSSION

1319A Beryllium

Annealing

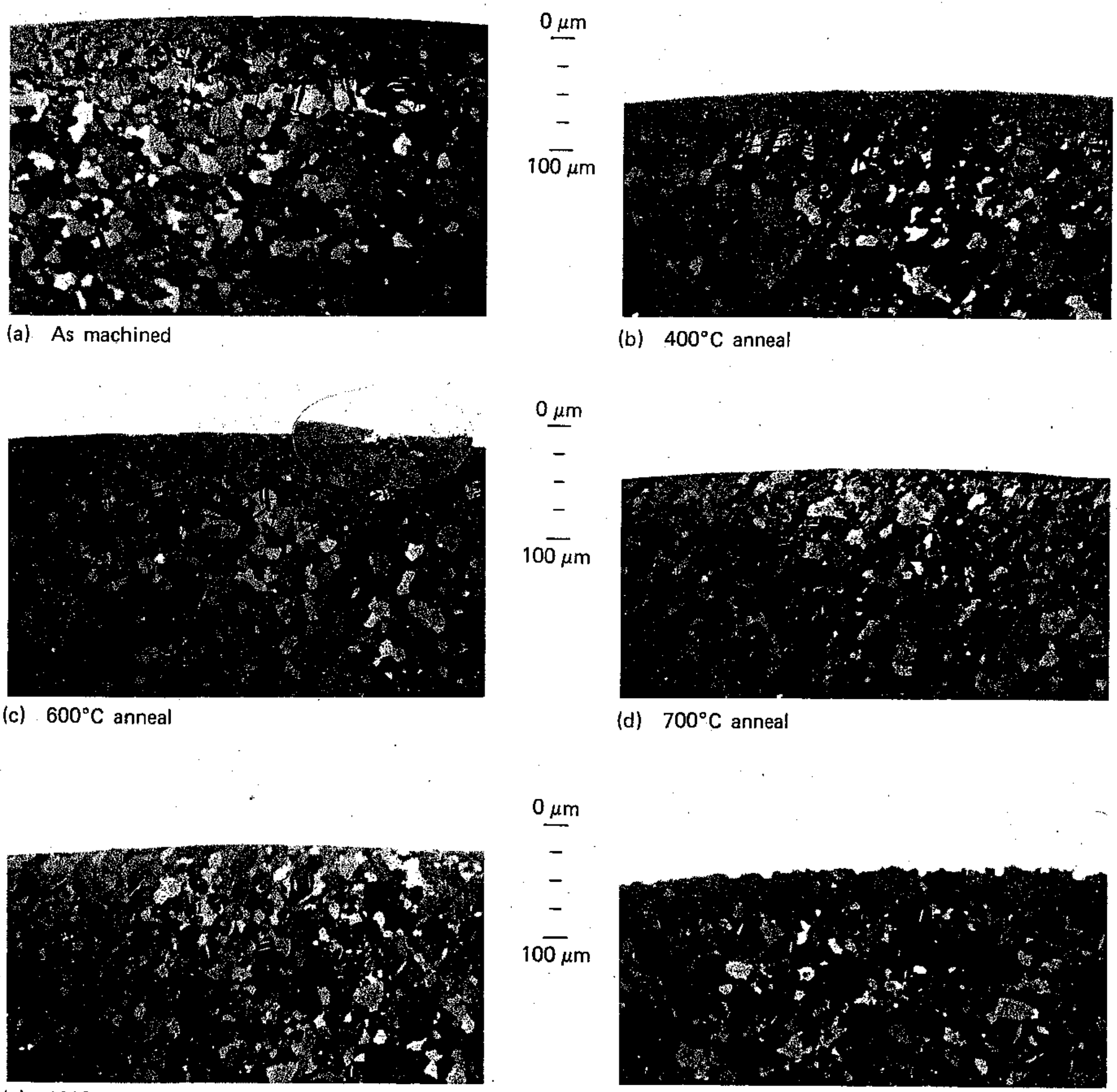
Table II summarizes the data obtained on all specimens in this study. Figure 1 shows the effects of machining damage and subsequent annealing for the 1319A beryllium. For this beryllium grade in the as-machined condition, the ultimate strength and the elongation at fracture were quite low. Annealing at temperatures up to 600 °C (1112 °F) did not significantly change these mechanical properties. However,

Table II. Transverse Tensile Properties of As-Machined, Annealed, and Etched Specimens

Grade	Condition	Ultimate Tensile Strength, MPa (ksi)	0.2 Pct-Yield Strength, MPa (ksi)	Yield Point, MPa (ksi)	Elongation, Pct
S200E	As-machined	263 (38.2)	216 (31.3)	—	0.7
	Annealed 400 °C	218 (31.6)	193 (28.0)	—	0.5
	500 °C	207 (30.0)	199 (28.8)	—	0.3
	600 °C	282 (40.9)	187 (27.2)	—	2.0
	700 °C	353 (51.2)	205 (29.7)	—	3.6
	800 °C	343 (49.8)	230 (33.3)	—	3.1
	900 °C	347 (50.4)	249 (36.1)	—	3.2
	1000 °C	342 (49.6)	239 (34.7)	—	3.3
	Etched 25 μm	323 (46.8)	209 (30.3)	—	2.8
	50 μm	337 (48.9)	220 (31.9)	—	2.8
	75 μm	333 (48.3)	216 (31.3)	—	3.1
1319A	As-machined	282 (40.8)	277 (40.2)	286 (41.4)	0.7
	Annealed 400 °C	249 (36.2)	245 (35.6)	257 (37.3)	0.4
	500 °C	270 (39.2)	—	—	0.2
	600 °C	242 (35.1)	—	—	0.2
	700 °C	375 (54.5)	257 (37.2)	264 (38.2)	3.1
	800 °C	381 (55.2)	235 (34.1)	—	3.8
	900 °C	388 (56.2)	238 (34.5)	—	4.0
	1000 °C	383 (55.6)	231 (33.5)	—	4.6
	Etched 25 μm	340 (49.3)	294 (42.7)	311 (45.2)	1.4
	50 μm	382 (55.4)	289 (41.9)	310 (45.0)	4.0
	75 μm	372 (53.9)	270 (39.2)	291 (42.2)	4.0

annealing at a temperature of 700 °C (1292 °F) did substantially increase both the ultimate strength and the elongation. Higher annealing temperatures—up to 1000 °C (1832 °F)—did not further improve these two properties. Thus the deleterious effect of the machining operation was removed by 1 h of annealing at temperatures of 700 °C (1292 °F) or above.

Annealing had a small but distinct effect on the 0.2 pct-yield strength and on the shape of the stress-strain curve. In the as-machined condition, there was an upper yield point, and the upper yield strength, the 0.2 pct-yield strength, and the ultimate strength were within just several megapascals of each other. The net result was a small plateau observed in the stress-strain curve of the as-machined specimen. The same behavior was observed after the 400 °C (752 °F) anneal. At annealing temperatures of 500 °C (932 °F) and 600 °C (1112 °F), the beryllium failed before reaching 0.2 pct yield and before exhibiting an upper yield point. However, the 700 °C (1292 °F) anneal removed the effect of machining damage, and, with higher ultimate strength and more ductility to fracture, the upper yield point was again observed. Raising the annealing temperature above 700 °C (1292 °F) caused a small decrease in 0.2



(a) As machined
(b) 400°C anneal
(c) 600°C anneal
(d) 700°C anneal
(e) 800°C anneal
(f) 1000°C anneal

Fig. 2—Effect of 1-h anneal at various temperatures on the microstructure of machining-damaged 1319A beryllium.

pct-yield strength and eliminated the upper yield point. The stress-strain curve observed at these temperatures was rounded and did not exhibit the small plateau in the yield-stress region. Beitscher *et al* reported very similar results on beryllium meeting the 1319A specification.[16] In Beitscher's study, the annealing treatment of 800 °C (1472 °F) for 2 h did alleviate the machining-damage effects, but it also eliminated a discontinuous yield point and resulted in a smooth, rounded stress-strain curve.

The changes in microstructure during the present, 1-h annealing treatments are shown in Fig. 2. The heavily damaged area of the as-machined condition is quite apparent in Fig. 2(*a*). It contains twinning to a depth of about 100 μm (0.004 in.) and a highly distorted grain structure, primarily in the first 25 μm (0.001 in.). Occasionally there is what appears to be a microcrack or possibly a smeared overlapped lip of beryllium (Fig. 2(*c*)). While these may not be cracks, they may still act as microstress concentrations, and, in that sense, they will have effects similar to those of a microcrack. Microcracks were not positively identified in the microstructure. The smeared overlaps were also observed in annealed tensile specimens in which the ductility had been restored.

As seen in Fig. 2, up through annealing temperatures of 600 °C (1112 °F), the appearance of the machining damage did not change; *i.e.*, the highly distorted and twinning microstructure showed no signs of recyrstallization. The anneal at 700 °C (1292 °F) nearly eliminated the machining damage in the first 50 μm, with twins remaining in the 50- to 100-μm (0.002- to 0.004-in.) region. The 700 °C (1292 °F) anneal also fully restored the tensile properties; thus apparently it is not necessary to remove all machining damage in order to recover the original mechanical properties.

The band of twinning was still present after annealing at 800 °C (1472 °F) but had disappeared after annealing at 1000 °C. There was some sublimation at 1000 °C (1832 °F) (Fig. 2(*f*)).

Microhardness surveys of the machining-damaged beryllium were quite revealing (Fig. 3). In the as-machined condition the microhardness was about 190 DPH toward the middle of the specimen, but, over the 100- to 125-μm (0.004- to 0.005-in.) machining-damaged area near the specimen surface the microhardness rose steeply to about 300 DPH. Thus, the highly worked and twinned region at the specimen surface has a much higher hardness than the original material. The pattern was similar after annealing at temperatures up to 600 °C (1112 °F), although the peak microhardness dropped somewhat. For specimens annealed at 700 °C (1292 °F), the hardness towards the middle of the specimen was still about 190 DPh and began to rise at about 100 to 125 μm (0.004 to 0.005 in.) from the specimen edge. However, in the 25- to 50-μm (0.001- to 0.002 in.) region immediately adjacent to the specimen edge, the microhardness dropped sharply to values actually a little lower than the original hardness. This corresponds precisely with the region of annealed microstructure at the specimen surface (Fig. 2) and the recovery of the tensile properties for these annealing temperatures (Fig. 1).

Annealing at 800 °C (1472 °F) resulted in lower microhardness than originally present for the first 25 to 50 μm (0.001 to 0.002 in.) from the specimen surface, but the microhardness returned to the original-material level at greater depth. Thus, annealing at 800 °C (1472 °F) alleviated all strain hardening from the machining damage and left the material immediately

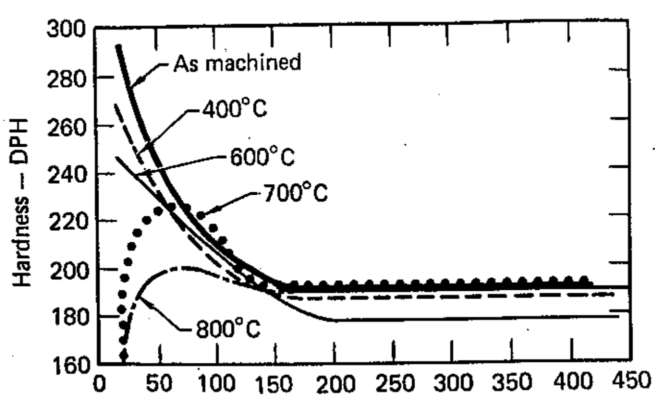


Fig. 3—Effect of various annealing temperatures on the microhardness of machining-damaged 1319A beryllium.

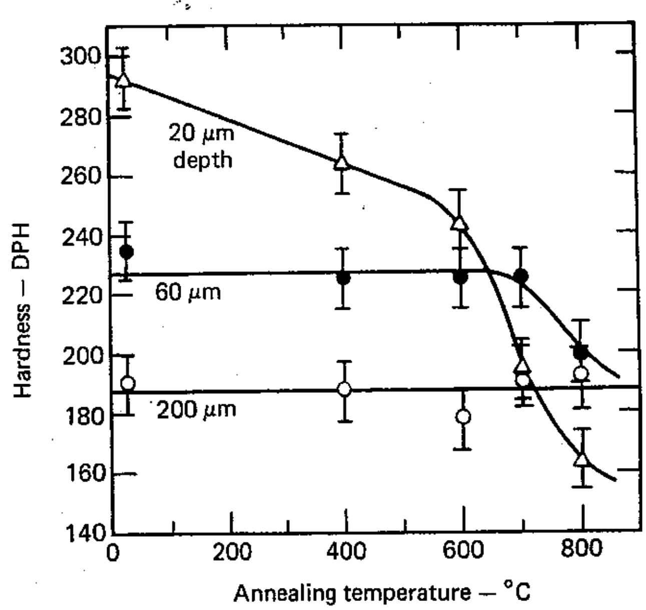


Fig. 4—Effect of annealing temperature on the microhardness at three specific distances from the specimen surface in machining-damaged 1319A beryllium.

adjacent to the specimen surface actually softer than in the original condition.

After the 1000 °C (1832 °F) anneal, the microhardness showed considerable scatter, ranging from about 140 to 220 DPH, although the lower values do tend to be near the specimen edge. This wide range of microhardness most likely results from both annealing and aging processes taking place simultaneously.

The importance of the concentration of cold working at the machined surface can be seen by examining the microhardness as a function of annealing temperature at constant depths. Figure 4 shows the microhardness at 20, 60, and 200 μm (0.0008, 0.0024, and 0.0080 in.). The 20-μm (0.008-in.) depth is about as close to the specimen surface as a reliable microhardness determination can be made. The 60-μm (0.0024-in.) depth is interesting because it is in the middle of the cold-worked band of twinning that persists during annealing, and the 200-μm (0.0080-in.) depth is well past the cold-worked surface layer.

At the 20-μm (0.0008-in.) depth, the microhardness decrease at annealing temperatures of 400 and 600 °C (725 and 1112 °F) indicates that some recovery from machining is taking place. At 700 °C (1292 °F) and above, the hardened surface layer is completely eliminated. At the 60-μm (0.0024-in.) depth, the microhardness is constant up through 700 °C (1292 °F),

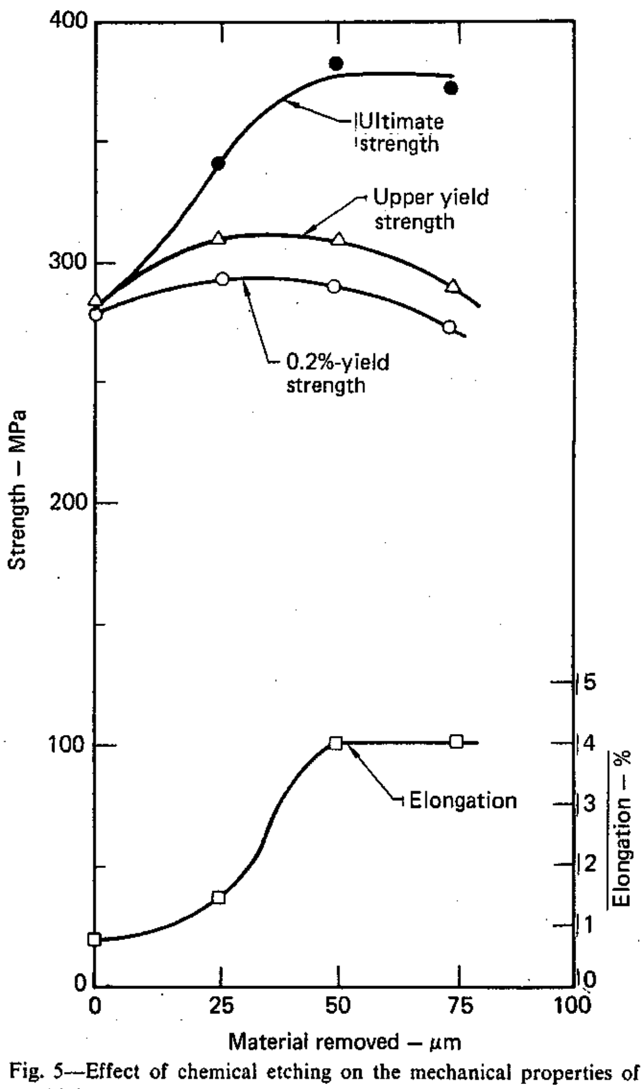


Fig. 5—Effect of chemical etching on the mechanical properties of machining-damaged 1319A beryllium.

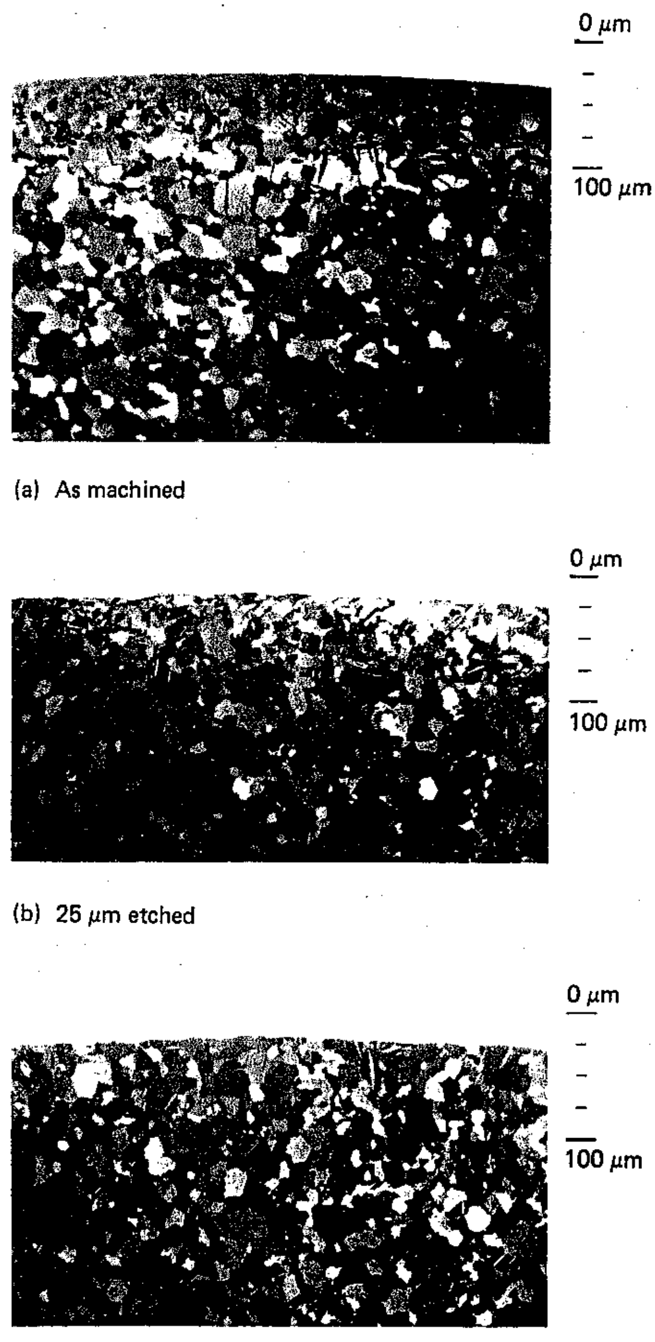


Fig. 6—Effect of chemical etching on the microstructure of machining-damaged 1319A beryllium.

indicating that the material here contains less distortion and stored energy to cause recovery and recrystallization. At the 200-μm (0.0080-in.) depth the microhardness shows no significant change up through the annealing temperature of 800 °C (1472 °F).

Etching

Etching just 25 μm from the surface of the machining-damaged beryllium improved both ultimate strength and elongation (Fig. 5). Etching 50 μm (0.002 in.) from the surface completely restored the tensile properties. Unlike the annealing, etching maintained the upper yield point while eliminating the effects of machining damage. There is a slight tendency for both the upper yield strength and 0.2 pct-yield strength to increase from etching of 25 and 50 μm (0.001 and 0.002 in.) and to decrease slightly from etching of 75 μm (0.003 in.), but the variations are within typical experimental scatter for beryllium.

After etching of 25 μm (0.001 in.), the machining damage in the microstructure is still very evident at the surface (Fig. 6), as would be expected from the original depth of the machining damage. After etching of 75 μm (0.003 in.), only a small amount of damage remained and the surface microstructure resembled the microstructure of the as-machined specimen at about 75 μm (0.003 in.) depth.

Figure 7 compares the microhardness of a 1319A tensile specimen on which 25 μm (0.001 in.) of material has been etched away with that of an as-machined specimen. When the etched-specimen data are offset 25 μm (0.001 in.) to compensate for the removed material, they follow those of the as-machined specimen within experimental scatter. Thus the etching is removing the most highly worked surface layer, but a worked layer still remains at the tensile specimen surface.

S200E Beryllium

Annealing

The effect of machining damage and subsequent annealing on the mechanical properties of S200E beryllium is shown in Table II and Fig. 8. The results are similar in many respects to those for the 1319A beryllium, but there are important differences. As in the 1319A beryllium, sufficient machining damage was introduced to substantially lower the elongation and ultimate strength. The 0.2 pct-yield strength was not affected. Annealing at 400 °C (725 °F) and 500 °C (932 °F) lowered the ultimate strength significantly and the 0.2 pct yield strength slightly.

The 600 °C (1112 °F) anneal resulted in some recovery of the ultimate strength and elongation, and the 700 °C (1292 °F) anneal fully restored the mechanical properties, including 0.2 pct-yield strength, to the original values. Anneals at 800 °C (1472 °F) and above do not cause any further change in the ultimate strength, yield strength, or elongation.

The changes in optical microstructure with annealing were similar to those observed for the 1319A beryllium (Fig. 9). Annealing up to 600 °C (1112 °F) produced no changes; twinning and highly distorted grains are still present. Annealing at 700 °C (1292 °F) recrystallized the microstructure in the first 50 μm (0.002 in.) from the specimen surface but left a band between approximately 50 and 100 μm (0.002 and 0.004 in.) that contains twins. Annealing temperatures of 800 °C (1472 °F) and above removed all evidence of machining damage.

Generally, the variation with depth of the microhardness in the as-machined S200E beryllium was quite

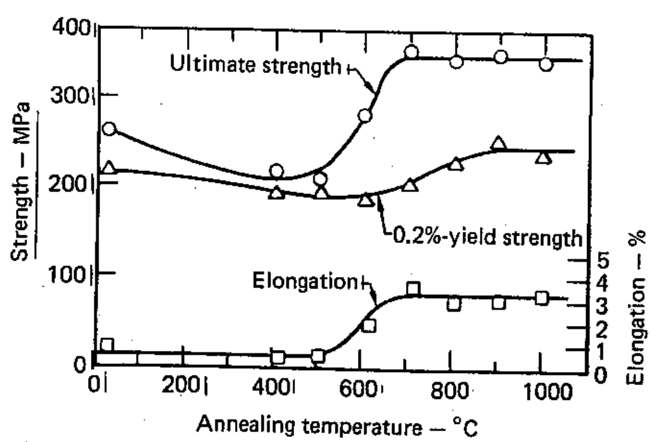


Fig. 7—Effect of etching 25 μm from the surface on the microhardness of 1319A beryllium.

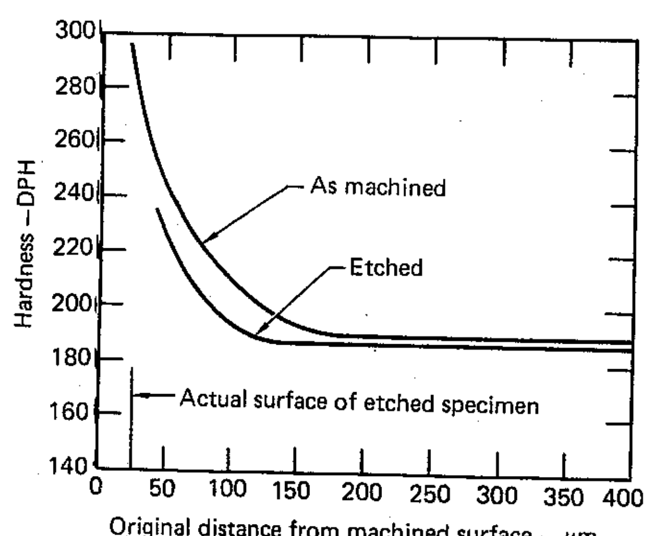


Fig. 8—Effect of 1-h anneal at various temperatures on the mechanical properties of machining-damaged S200E beryllium.

similar to that observed for the 1319A beryllium (Fig. 10). The microhardness below the machining damage, near the center of the specimen, was in the 150 to 180 DPH range. In the machining-damaged region (approximately the first 100 to 125 μm (0.004 to 0.005 in.)), the microhardness rose steeply in a continuous manner to about 280 DPH.

In this S200E grade, the microhardness responded much differently than in the 1319A beryllium. Annealing at 400 °C (752 °F) introduced a substantial variation in microhardness, with the values in the machining-damaged region ranging from 200 to 250 DPH. Beneath the machining-damaged region, the hardness ranged from 185 to 230 DPH. As was seen in Figs. 8 and 9 this anneal and the 500 °C (932 °F) anneal produced no changes in the optical microstructure but did significantly decrease the ultimate strength and slightly decrease the 0.2 pct-yield strength.

After the 600 °C (1112 °F) anneal, the microhardness away from the machining-damaged region was similar to that in the original material and rose steeply in a continuous fashion to about 230 DPH in the machining-damaged region. The 700 °C (1292 °F) anneal substantially lowered the microhardness near the specimen

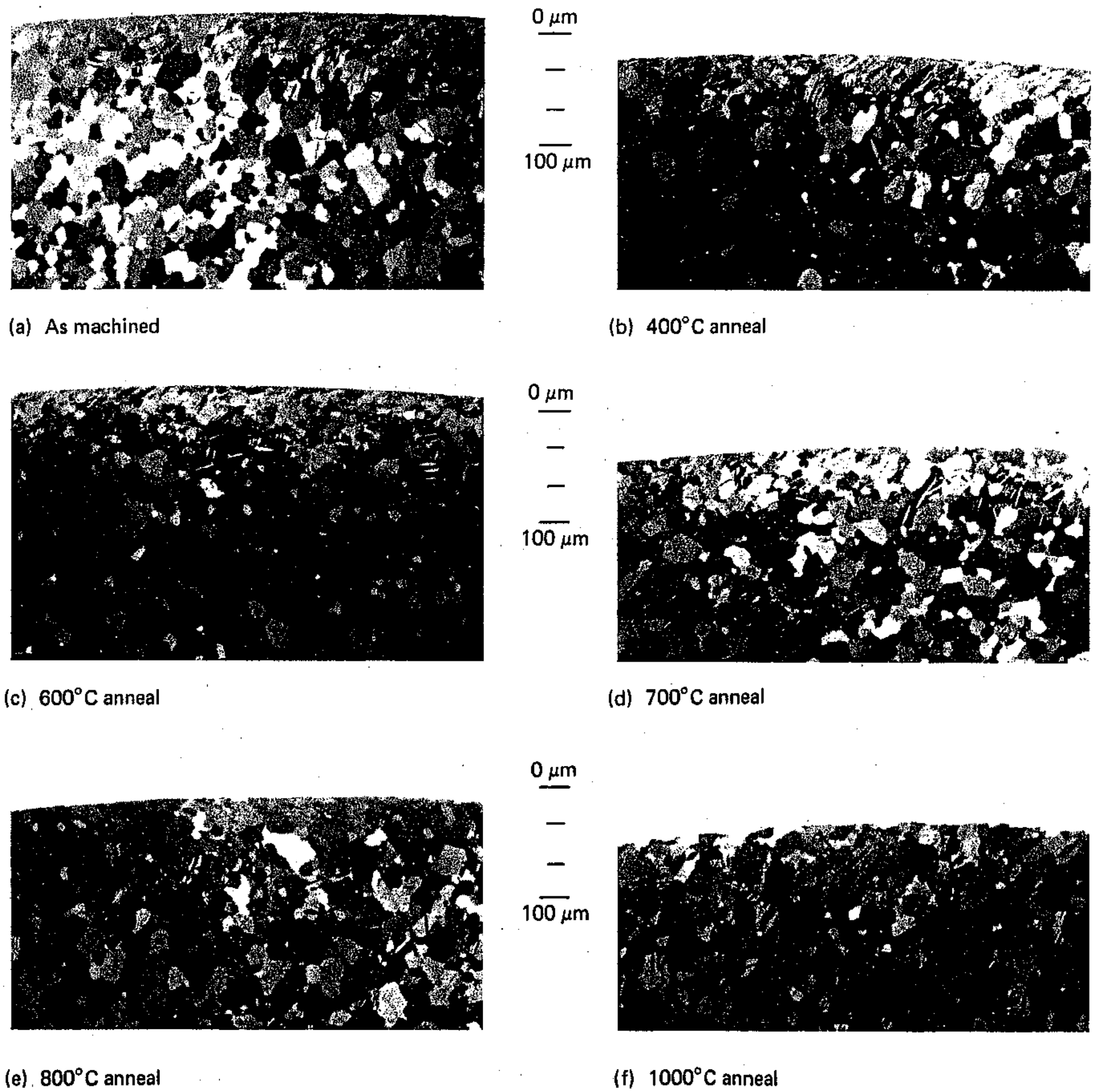


Fig. 9—Effect of 1-h anneal at various temperatures on the microstructure of machining-damaged S200E beryllium.

edge; the microhardness rose in the region between about 25 and 75 μm (0.001 and 0.003 in.) and then decreased to the original levels at greater depths. This variation correlates with the optically observed microstructure showing machining damage in the 25- to 75-μm (0.001to 0.003-in.) range and an annealed microstructure at the specimen surface. The 800 °C anneal eliminated the machining damage completely, resulting in microhardness values similar to those in the original material, except at the specimen edge, where the microhardness decreased rapidly in a manner similar to that resulting from the 700 °C (1292 °F) anneal.

As occurred in the 1319A beryllium, annealing at 1000 °C (1832 °F) produced substantial changes in the microhardness of the S200E beryllium. Unlike in the 1319A beryllium, though, the microhardness of the S200E beryllium after this anneal did not show scatter, but was raised uniformly from the 800 °C (1425 °F) anneal levels. This resulted in a soft area immediately adjacent to the surface, high values in the 50- to 100-μm (0.002to 0.004-in.) range, and a lower, constant values toward the center of the specimen. The large but uniform change results from aging phenomena taking place in this microalloyed grade of beryllium. Determination of the precise mechanism for the raising of the microhardness is beyond the scope of the present study, but the essential point is that, in some grades of beryllium, annealing to eliminate machining damage can cause many other changes.

The combined effects of the cold-working and aging phenomena is particularly apparent in comparing the microhardnesses of the annealed specimens at 20, 60, and 200 μm (0.0008, 0.0024 and 0.0080 in.) (Fig. 11). In the highly worked 20-μm (0.0008 in.) region, the dominant effect is the annealing of the machining damage, resulting in some decrease in hardness at 400 °C (725 °F) and 600 °C (1112 °F) and a substantial drop at higher temperatures up through 800 °C (1472 °F). At both 60 and 200 μm (0.0024 and 0.0080 in.) the hardness increases noticeably at 400 °C (752 °F). At

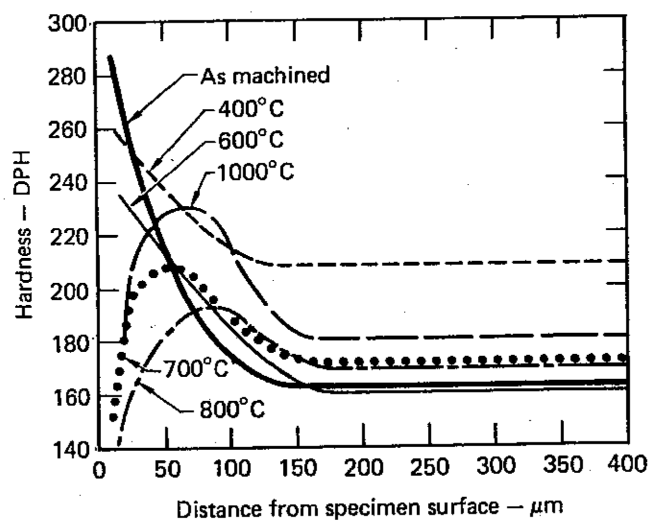


Fig. 10—Effect of various annealing temperatures on the microhardness of machining-damaged S200E beryllium.

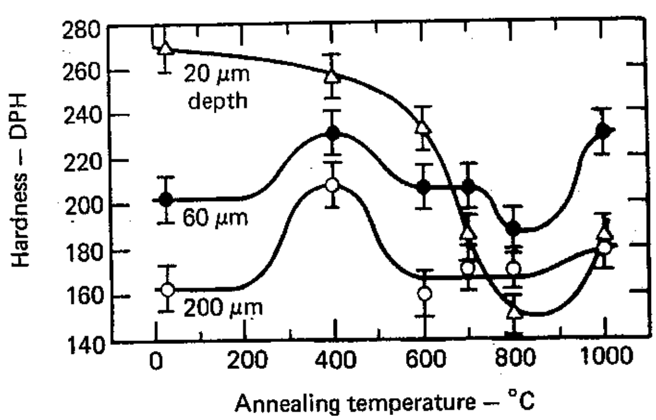


Fig. 11—Effect of annealing temperature on the microhardness at three specific distances from the specimen surface in machining-damaged S200E beryllium.

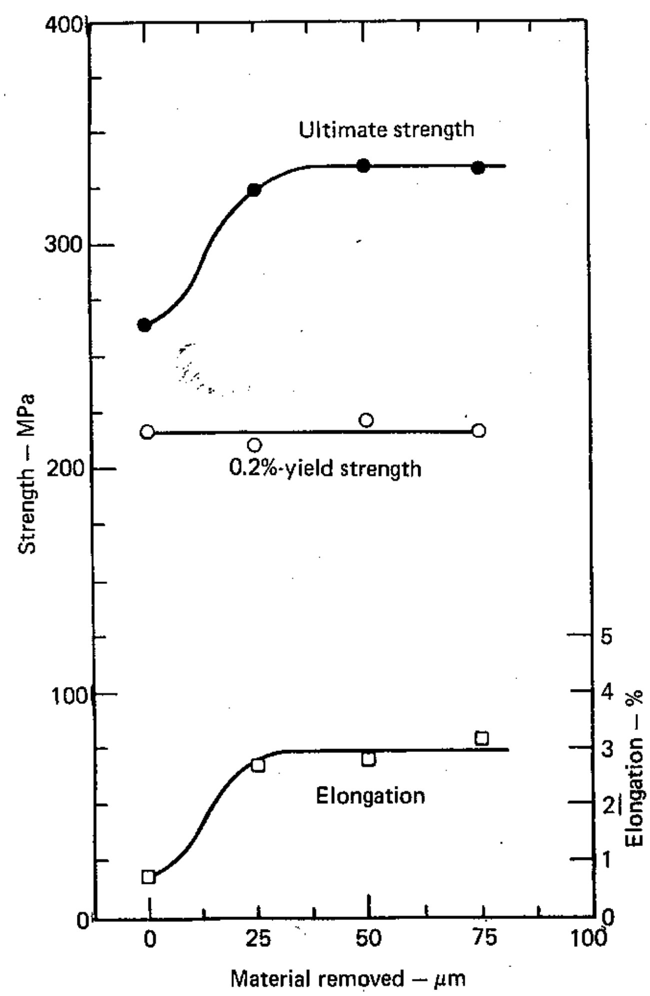


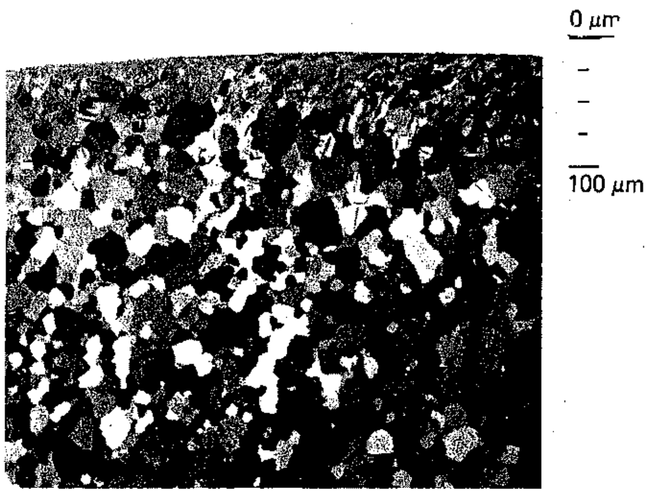
Fig. 12—Effect of chemical etching on the mechanical properties of machining-damaged S200E beryllium.

1000 °C (1832 °F) the increase in hardness at all three distances is clearly evident, although the tensile properties were no different at 1000 °C (1832 °F) than at 800 °C (1472 °F).

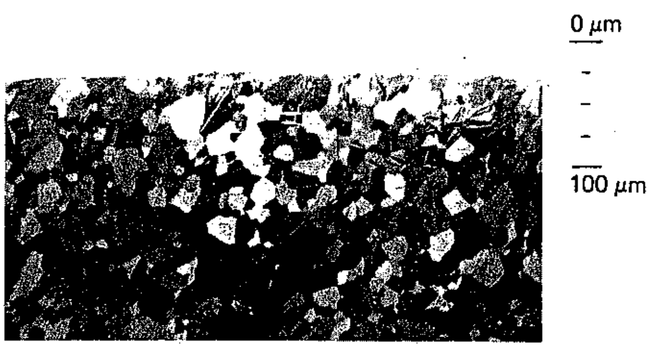
Etching

Chemical removal of just 25 μm (0.001 in.) from the surface of the S200E specimens restored the ultimate strength and elongation (Fig. 12). Etching 50 and 75 μm (0.002 and 0.003 in.) did not further improve the mechanical properties. The yield strength did not change at any of the three levels of etching.

As in the microstructure of the 1319A beryllium, the S200E grade appears to still contain a significant amount of machining damage after chemical removal of 25 μm (0.001 in.) of material (Fig. 13). The remaining damage is primarily twinning to depth of 25 to 50 μm (0.001 to 0.002 in.) and some grain distortion in the first 25 μm (0.001 in.). The presence of this machining damage did not lower the ultimate strength or elongation from the certified values shown in Table I.



(a) As machined



(b) 25 μm etched

(c) 75 μm etched

Fig. 13—Effect of chemical etching on the microstructure of machining-damaged S200E beryllium.

SUMMARY AND CONCLUSIONS

The machining damage introduced in this study greatly decreased ultimate strength and ductility but did not affect yield strength. For both grades of beryllium studied, annealing temperatures of at least 700 °C (1292 °F) for 1 h were needed to completely restore the original mechanical properties. However, annealing at temperatures both below and above 700 °C (1292 °F) also changed other mechanical properties. The specific property changes noted were the disappearance of the yield-point phenomenon, decreases in 0.2 pct-yield and ultimate strengths, and substantial changes in hardness. The existence and magnitude of these other changes depended on the annealing temperature and the grade of beryllium. Thus, if annealing is going to be employed to eliminate the effects of machining damage, it is important to monitor other mechanical properties and ensure that they do not change significantly.

For the extensive machining damage introduced in this study, the original ultimate strength and ductility were restored by chemical removal of only 25 to 50 μm (0.001 to 0.002 in.) of material. The 0.2 pct-yield strength and, when present, the upper yield point were not noticeably affected by the presence of machining damage or by the chemical etching.

In both annealing and etching, mechanical properties were restored without complete removal of all microscopic evidence of machining damage. In annealing, it was sufficient to eliminate the machining damage in the first 25 to 50 μm (0.001 to 0.002 in.), leaving a damaged layer between 50 and 100 μm (0.002 and 0.004 in.) from the specimen edge. In chemical etching, it was necessary to remove only the first 25 to 50 μm (0.001 to 0.002 in.) of the machining-damaged material to restore properties. However, etching left machining-damaged material at the surface, and the damage extended inward approximately 25 to 50 μm (0.001 to 0.002 in.).

Thus, the deleterious effect of machining lies only within a very thin layer of material at the surface. While twinning and microcracks are usually suggested as the critical factors in machining damage, residual stresses and a layer of textured material at the surface have also been suggested. Microcracks were not observed other than as a result of possibly folded or smeared surface material, and they were present in both the as-machined condition and the annealed material that had restored mechanical properties. Annealing restored the mechanical properties, and while annealing could eliminate internal cracks,[17] it would probably not close surface cracks. Thus surface microcracks could not have contributed to the loss of physical properties in this study.

Researchers have shown that twins can act as substantial stress risers in beryllium,[4] but in this study twinned microstructures were observed at the surface of tensile specimens that exhibited optimum mechanical properties. While the density of twins was not high in such cases, this does demonstrate that twinning by itself does not necessarily degrade mechanical properties.

Mechanically working beryllium can easily produce strong textures. For example rolling beryllium sheet can quickly produce a (0001) texture from four to ten times random. If machining beryllium produces an equivalent situation, the result could be a basal surface texture with the prism planes tending to be approximately perpendicular to the tensile axis. Such a surface layer would be ideally oriented for prism-plane cleavage. In the present study, the optical appearance of the microstructure, the presence of a very high microhardness near the surface, and the response of the surface regions to annealing all indicate a highly worked surface layer. This evidence indicates that more serious consideration should be given to the possibility that a strain-hardened, oriented surface layer is a basic cause of low mechanical properties in machining-damaged beryllium.

This possible mechanism would have two consequences. First, if annealing is used to eliminate machining damage, simple metallography and tensile tests alone will not reveal the optimum time and temperature of anneal. Such evaluations could be used as a guide, but the results would have to be related to the extent of strain hardening and surface texture caused by the actual machining procedure being used to manufacture the part. Second, microhardness at the surface might be used in some situations to ascertain the extent of machining damage. While it is not a nondestructive test and there is some risk in initiating microcracks at the indent, it is less destructive than cutting up the entire part to remove metallographic samples.

ACKNOWLEDGMENTS

R. E. Burtis' skill and advice was very useful in the key step of introducing the machining damage in a consistent controlled manner. F. A. Dishong performed all the etching and annealing experiments; G. D. Estes carried out the mechanical tests; and R. P. Kershaw carried out the metallography, carefully bringing out the highly distorted edge microstructure. The author appreciates technical discussions with J. B. Mitchell.

This work was performed under the auspices of the U.S. Department of Energy by the Lawrence Livermore Laboratory under contract number W-7405-Eng-48.

REFERENCES

1. A. R. Kaufman, P. Gordon, and D. W. Lillie: *ASM Trans.*, 1950, vol. 42, pp. 785–844.
2. W. V. Ward, M. I. Jacobson, and C. O. Matthews: *ASM Trans.*, 1961, vol. 54, pp. 84–95.
3. M. I. Jacobson, F. M. Almeter, and E. C. Burke: *ASM Trans.*, 1962, vol. 55, pp. 492–504.
4. W. Bonfield, J. A. Sartell, and C. H. Li: *AIME Trans.*, 1963, vol. 227, pp. 669–73.
5. N. A. Hill: *The Metallurgy of Beryllium*, p. 84, Chapman and Hall, London, 1963.
6. M. Herman and G. E. Spangler: *The Metallurgy of Beryllium*, p. 75, Chapman and Hall, London, 1963.
7. V. M. Hovis: Rept. No. Y-1551, Union Carbide, Oak Ridge Plant, 1966.
8. S. Beitscher: Rept. No. RPF-1205, Rockwell International Rocky Flats Plant, 1968.
9. J. L. Frankey and D. R. Floyd: Rept. No. RFP-910, Rockwell International Rocky Flats Plant, 1968.
10. J. A. Gurklis: Rept. No. MCIC-72-03, Metals and Ceramics Information Center, 1972.
11. R. L. Riegel, E. L. Childs, and K. R. Souply: Rept. No. RFP-2174, Rockwell International Rocky Flats Plant, 1974.
12. W. Taylor and J. S. White: *Int. J. Met. Sci.*, 1974, vol. 9, pp. 569–75.
13. B. King: Rept. No. AFML-TR-76-88, Air Force Materials Laboratory, Wright-Patterson Air Force Base, 1976.
14. J. E. Hanafee, J. W. Hughes, Jr., and S. A. McInturff: Rept. No. UCID-17578, Lawrence Livermore Laboratory, 1977.
15. J. E. Hanafee: Rept. No. UCRL-52287, Lawrence Livermore Laboratory, 1977.
16. S. Beitscher, J. F. Lapes, W. M. Leslie, J. R. Luchow, and R. L. Riegel: Rept. No. 2838, Rockwell International, Rocky Flats Plant, 1979.
17. G. I. Turner, R. A. Lane, and R. A. Lancaster: *Fourth International Conference on Beryllium*, paper 12, The Metals Society, London, 1977.
18. R. E. Evans: *Fourth International Conference on Beryllium*, paper 37, The Metals Society, London, 1977.
19. Materials Advisory Board: Rept. No. MAB-205-M, National Academy of Sciences, National Research Council, Washington, D.C., 1966.
20. J. E. Hanafee: Specification MEL 76-001319A, Lawrence Livermore Laboratory, 1976.