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EFFECT OF MOLYBDENUM VARIATIONS ON THE CREEP-RUPTURE PROPERTIES OF A COMPLEX NICKEL-BASE ALLOY

by

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The role of molybdenum in heat-resistant Cr - Ni or Cr - Ni - Co alloys containing Ti + Al as main strengthening elements is not understood. Alloys of this type usually contain up to 4-percent or about 10-percent Mo. It is impossible to determine from published information to what degree the Mo in commercial alloys represents optimum amounts. Furthermore, intermediate amounts between 4 and 10 percent is reputed in at least one case (Mo between the 4 and 10 percent in Waspaloy and M252 alloys) to have had adverse effects.

In this investigation being reported, the effect of Mo was systematically studied between 0 and 12 percent in steps of 2 percent in a 20 Cr - 55 Ni - 15 Co - 3 Ti - 3 Al base alloy. This alloy with 4 percent Mo is known most generally as Udimet 500 alloy. The heats were melted in vacuum and weighed about 11 pounds. Additions were made to give a boron content of 0.0060 percent and a zirconium content of 0.05 percent. These were rolled to bar stock and given the standard double solution treatment at 2150° and 1975°F and double-age at 1550° and 1400°F for Udimet 500 alloy. Two creep-rupture tests were carried out to approximately determine the rupture strength for 100 hours at 1600°F as a function of the Mo content. Additional stock was furnished to the Climax Molybdenum Company for other tests, mainly oxidation studies.

SUMMARY AND CONCLUSIONS

The data were sufficient to indicate that the optimum amount of Mo was in the range of 2 to 4 percent. A marked improvement in 100-hour rupture strength and ductility at 1600°F resulted from the addition of 2 percent Mo. Rupture strength then fell off with Mo content above 4 percent up to 12 percent. Ductility in the rupture tests decreased to a minimum between 6 and 8 percent of Mo and the increased markedly at 10 and 12 percent. Some difficulty was encountered

from center cracking during the hot rolling of the ingots containing 10 and 12 percent of Mo. These alloys were also brittle at room temperature in the aged condition.

A limited study of structures indicated that the lower strength and high ductility in rupture tests and brittleness at room temperature at 10 and 12 percent of Mo was associated with extensive sigma phase. The reduced amount of γ' precipitate in these alloys probably was a major contributing factor to the lower strength. No definite information was obtained to indicate the cause of fall-off in strength between 4 and 8 percent. The reduced ductility in this range of Mo contents may be the reason why alloys with these amounts of Mo are not used.

The results and limitations of the investigation are discussed in some detail in the report. The major factor to recognize is the probability that optimum Mo contents vary with the Ti + Al contents in alloys. Certainly lower Ti + Al alloys such as M252 with 10-percent Mo cannot be subject to the sigma phase reaction encountered in the experimental alloy used for this investigation.

EXPERIMENTAL PROCEDURES

General Considerations

The techniques detailed in the following section represent the final procedures adopted for this investigation. The necessity for producing larger ingots (11 pounds) than those which had been made previously (8 pounds) in this laboratory introduced several inter-related problems which led to modification of the existing techniques.

The dimensions of the new ingot were limited by the maximum roll spacing available in the rolling mill because the ingots were to be direct rolled to bar stock. As a result, the increased weight of the ingot had to be largely derived from an increase in the length of the mold, with only a small amount of it being achieved through an increase in mold diameter. This led to problems which had not previously been encountered in mold design involving a small diameter center pipe

in the ingot. Several time consuming modifications of the mold were necessary before the piping was minimized to the extent that satisfactory ingots could be produced.

During trial rolling of some of the first ingots produced, it was thought that the larger diameter of the ingots was increasing the severity of the rolling operation on the metal, since surface cracking was considerably more serious than had previously been experienced. For this reason it was decided to raise the aim boron level from the 0.0030 percent originally established to 0.0060 percent. This materially reduced cracking during rolling. Subsequent experience, however, demonstrated that the increased cracking was limited to those heats with more than 6-percent Mo.

Finally, it was discovered that the three highest molybdenum heats were subject to severe internal rupturing under the standard rolling schedules. For this reason the rolling temperature and amount of reduction per pass had to be increased.

Melting

The alloys were induction melted using alumina crucibles in the University of Michigan vacuum-melting unit shown in Figure 1. Pressure before melting down was less than 5 microns as measured by both Stokes and thermocouple gauges; the leak rate was 5 microns per minute or less. No gases were purposely added during the melting cycles. Temperatures were read using an optical pyrometer and pouring was at 150-200°F above the measured freeze point for each alloy.

The aim analysis of the alloy series was as follows:

1. Constant aim for all heats:

(Weight Percent)

C	Mn	Si	Cr	Ti	Al	B	Zr
0.08	0.10	0.20	19.0	3.0	3.0	0.0060	0.05

2. Varied among heats maintaining a constant Ni/Co ratio:

Heat	Mo	Ni	Co
1	0	55.5	18.9
2	2	54.0	18.4
3	4	52.5	17.9
4	6	51.0	17.4
5	8	49.6	16.8
6	10	48.1	16.3
7	12	46.6	15.8

Laboratory-grade electrolytic nickel, chromium, cobalt and manganese melting stock was used. The molybdenum was arc-melted, low-carbon stock; the titanium was Ti 55A bar stock; the aluminum was 99.99 percent purity ingot; the silicon was 99.9 percent purified powder; zirconium was added as sponge; boron as nickel-boron master alloy; and carbon as powdered graphite.

The ingots were cast in a copper mold. A drawing of the mold design is shown in Figure 2. The design includes a relatively thin-walled upper portion with a larger bell at the bottom to serve as a heat-sink at that point. A heavy copper plate is bolted to the bottom of the mold to promote progressive solidification from the bottom. A hot top made from low density insulating brick is set on top of the mold to feed the ingot during solidification. The relevant dimensions of the design are indicated on the drawing.

Hot Working and Heat Treatment

The alloys were hot worked to 1-inch round-corner square bar stock using a two-high mill with grooved rolls. The ingots were cut in half and heated for 2 hours at the rolling temperature prior to the start of rolling. They were then reheated for ten minutes after every pass through the mill.

The ingots with 0, 2, 4, and 6 percent molybdenum were rolled using the techniques developed for rolling standard Udimet 500 alloy. The 2 hour preheat was at 2150°F followed by rolling at 2150°F using reductions of about 7 - 10 percent. For the three higher molybdenum heats, it was necessary to reduce the rolling temperature by 100° to 150°F and roll with a corresponding decrease in reduction per pass to about 5 - 7 percent.

Following the rolling, sufficient bar stock for four rupture specimens and

a four-inch length of stock for the proposed oxidation studies were heat treated as follows:

1. 4 hours, 2150°F, air cool;
2. 4 hours, 1975°F, air cool;
3. 24 hours, 1550°F, air cool; and
4. 16 hours, 1400°F, air cool.

In view of the variation in rolling conditions represented among the heats, four hours at 2150°F was used to minimize any residual differences from the fabricating.

Chemical Analysis

The chemical analyses were obtained from the Metallurgical Products Department of the General Electric Company by the Climax-Molybdenum Company. Three of the heats were completely analyzed by wet chemical methods. The Mo, Co, Cr, and Ti in the other four heats were then analyzed by X-ray techniques.

Creep-Rupture Testing

The 100-hour rupture strength at 1600°F was estimated for each alloy with two creep-rupture tests. The tests were run on 0.250-inch diameter samples with a 1-inch gage length. The samples were held at 1600°F for 4 hours before application of the stress. Samples for the five lower molybdenum heats were machined from the fully heat treated bar stock. The materials with 10 and 12 percent of molybdenum could not be cut in this condition, however, and were machined from the stock after the 1975°F treatment. The finished samples were then aged and tested.

Creep data were taken for every test with an optical extensometer system having a sensitivity of 0.00001 inch.

RESULTS AND DISCUSSION

Chemical Composition

The chemical analyses, as obtained from the Metallurgical Products Department of the General Electric Company by the Climax Molybdenum Company,

are given by Table I.

Since melting practice was constant for all of the heats, only three heats, the 0, 6, and 12 percent molybdenum alloys, were completely analyzed by wet chemical methods. The results of these analyses were then used to calibrate for X-ray analysis of all seven heats for the elements Mo, Co, Cr, and Ti.

The chemistry during melting was generally in good control. The aim molybdenum values were closely approximated and the cobalt variations associated with them were equally good. The boron level was slightly below the aim amount of 0.0060 percent but was fairly consistent among the three heats checked. The zirconium content was not determined on an absolute basis due to lack of acceptable standards. The intensity of its radiations indicated that the level was constant among the heats.

Hot-Workability

Relatively little difficulty was encountered in rolling the heats with 0, 2, 4, and 6 percent of Mo after the boron addition was increased to 0.0060 percent although some corner cracking occurred.

The heat with 12 percent of Mo was subject to internal rupturing when rolled at 2150°F with reduction of 7 to 10 percent per pass, the conditions used for the lower Mo heats. The temperature was reduced to 2000°F and the reduction per pass was reduced to about 5 percent. The heats with 8 and 10 percent of Mo were rolled first at 2000°F. After about one-third of the reduction of the ingot the temperature was raised to 2050°F. Fewer corner cracks were encountered at these lower rolling temperatures although those which did form were more difficult to control by grinding and were deeper than in the heats with lower Mo.

Heat-Treatment

Following the hot working, a three-inch length of bar stock from each heat was heat treated for rupture tests. At the same time a four-inch length was heat treated and forward to the Climax Molybdenum Company for proposed oxidation tests.

The heat treatment on these pieces included the following:

4 hours at 2150°F, air cooled
4 hours at 1975°F, air cooled
24 hours at 1550°F, air cooled
16 hours at 1400°F, air cooled

The bars heat treated for rupture specimens were then quartered lengthwise to provide four blanks suitable for machining into specimens. The material from the heats with 10 and 12 percent of Mo cracked severely during this operation. Metallographic examination indicated that these two heats contained a large amount of an acicular phase after aging. Another set of bars for these two heats was given the 2150° and 1975°F treatments and successfully machined to specimens in this condition. These machined specimens were then given the two aging treatments.

Creep-Rupture Properties

The results of the rupture tests at 1600°F are given in Table I. The rupture times from the two tests for each Mo level were plotted to give approximate stress-rupture time curves (Figure 3). The approximate stress for rupture in 100 hours indicated for each alloy by these curves is given in Table II.

The rupture strengths for 100 hours at 1600°F show a maximum in the range of 2 to 4 percent of molybdenum (Figure 4a). Additions in this range raised the rupture strength about 6500 psi from that of the alloy with no Mo. The strength appeared to drop off fairly regularly up to 8-percent Mo to about the same level as for the Mo free alloy. There was an abrupt decrease between 8 and 10 percent of Mo to a level about 4000 psi below the Mo free alloy for the heats with 10 and 12 percent.

The addition of 2 to 4 percent of Mo was very beneficial to ductility in the rupture tests (Table II and Figure 4b). A somewhat low minimum in ductility apparently occurs between 6 and 8 percent of Mo followed by a very marked increase for 10 and 12 percent.

It should be clearly recognized that the two tests do not completely fix the

100-hour rupture strengths. The difference in slope (Figure 3) of the stress-rupture time curves for the heats with 6 and 8 percent of Mo suggests that these two in particular are open to some question. On the other hand it is also evident that additional tests fixing the strength with more certainty would not significantly change the influence of Mo as indicated by Figure 4. Due to the uncertainty of the 100-hour rupture strengths an average curve has been drawn through the data in Figure 4. Further testing of the particular heats involved as well as of additional heats would be required to establish whether the maximum does exist at 2-percent Mo as the limited data suggest. The breaks shown in the curves between 8 and 10-percent Mo for both strength and ductility is believed to be real and due to the appearance of sigma phase as discussed in the next section.

The creep curves obtained from the rupture tests are shown in Figures 7 through 13. Due to the variation in stresses it is somewhat difficult to determine the influence of Mo on creep characteristics. It is evident, however, that there was a marked increase in creep resistance from the addition of 2-percent Mo. The heat with 4-percent Mo was less creep resistant than the 2-percent. There was not much further decrease in creep resistance for 6 and 8 percent of Mo and then a marked decrease at 10 and 12 percent.

Structural Studies

The cracking encountered when the alloys with the larger additions of Mo were machined in the fully heat treated condition was unexpected. A metallographic examination disclosed an acicular phase present after aging. The amount present in the alloy with 8-percent Mo was very small and was difficult to locate in the microstructure. Small amounts could be found in localized areas associated with carbides as illustrated by the electron micrographs of Figure 5(a). While this figure suggests that aging at 1400°F reduced the amount of acicular sigma phase it should be recognized that it was so difficult to find evidence of the phase in this alloy that the electron micrographs cannot be typical.

The amount of the sigma phase was markedly increased at 10 Mo (Figure 5(b)). It was also increased further by 12 Mo (Figure 6) although this is not particularly evident in the electron micrographs. These latter micrographs also show that the major amount of the sigma phase developed during aging at 1550°F although some indication of an acicular phase remains in the material as solution treated (Figure 6(a)). Evidently precipitation of this phase was not completely suppressed by solution treating.

Certain other features of the structures are suggestive of possible important effects:

1. The amount of γ' phase appears to have been reduced in alloys with 10 and 12 percent of Mo. Normally the matrix is completely occupied by the γ' phase as is shown in the structure of the 8-Mo material (Figure 5(a)) after double aging. Those heats containing more sigma phase show less γ' phase. The γ' is the small particles in the background matrix as shown by the electron micrographs.

2. The γ' particles are small cubes. Normally these are spheres in the 4-Mo base alloy.

3. The treatment at 1975°F apparently was not high enough to completely dissolve all the γ'. The samples of the material with 0, 2, and 4 percent of Mo were examined under the optical microscope. There seemed to be segregated areas of γ' which probably had not been dissolved at 1975°F. It is known that 1975°F is marginal for completely dissolving the γ' in 3Ti-3Al alloy and it is not too surprising to find undissolved γ' in the alloys.

4. Certain of the electron micrographs of Figures 5 and 6, as well as the results of visual study of the structures, suggest that the acicular phase is actually in the form of plates.

An extraction was carried out by J. A. Amy at the University on the 10 Mo alloy using hot aqua-regia to separate out the acicular phase. The diffraction data used to make the identification are given in Table III.

The acicular sigma phase would be expected to embrittle the alloy at room temperature as was observed for 10 and 12 Mo. The solution treatments used did not completely suppress the acicular phase. The material was, however, certainly more ductile than after aging.

Alloys with sigma phase are generally comparatively weak and ductile at high temperatures. It seems probable, however, that the reduction in γ' was probably the most important cause for reduced strength in the alloys with 10 and 12 Mo.

The microstructures suggest that the γ' reaction was affected by 8, 10, and 12 Mo. This was evident in the cubic rather than spherical nature of the γ' particles. Certainly when sigma phase formed the amount of γ' was reduced.

DISCUSSION

The results indicate 2 to 4 percent of Mo imparted the maximum improvement in rupture properties for the alloy investigated in 100 hours at 1600°F. The information obtained was not sufficient to explain the mechanism of improvement for these additions of Mo or the fall-off in properties when the Mo was increased to 6 and 8 percent. The identification of sigma phase in the alloys with Mo contents of 10 and 12 percent accounts for the reduced strength of those alloys. Sigma phase itself is weak and ductile at high temperatures. It is probably more important, however, that the amount of γ' phase in the matrix was reduced in the presence of the sigma phase, an effect which would be expected to markedly reduce strength.

The 4-percent Mo standard heat had properties which were as good as the best from an extensive series of heats made in this laboratory to study melting practice and trace element effects. The trends of the effect of Mo found were therefore not due to poor properties in the basic heat.

It is noteworthy that ductility in the rupture tests became a minimum between 6 and 8 percent Mo. Commercial Cr-Ni alloys precipitation hardened with Ti + Al

contain either up to 4-percent or 10-percent Mo. Intermediate amounts between 4 and 10 percent are not used. It has furthermore been reported that intermediate amounts of Mo resulted in undue brittleness in an alloy based on the Waspaloy-M252 base composition. The results of this investigation suggest that the observed embrittlement may be characteristic of 6 to 8 percent of Mo. No indication was found in the alloy studied to indicate that sigma phase caused the reduced ductility. The good properties of M252 and René 41 alloys indicate that sigma phase does not occur with 10-percent Mo for the lower Ti + Al contents in these alloys. Possibly a recovery of rupture strength and ductility occurs as Mo is increased from 6 - 8 percent to 10 percent in the absence of sigma phase.

It is important to recognize that the effectiveness of Mo could vary to a considerable extent with several factors:

1. The level of Ti + Al may shift optimum amounts of Mo. For instance, the smaller amounts of γ' in lower Ti + Al alloys may allow Mo to be more effective though solution strengthening or the precipitation of carbides. The appearance of sigma phase when Mo and Ti + Al contents are high has previously been discussed.

2. Experience at the University suggests that a major effect of 10-percent Mo is to retard the formation and growth of microcracks in the grain boundaries, the mechanism by which rupture occurs in the alloy. B and B + Zr, however, operate in the same manner. It was noted that additions of 2 to 4 percent Mo markedly increased ductility in rupture tests for the alloy studied. Thus the optimum Mo content might be quite sensitive to the amounts of B and Zr in the alloy due to the competitive similarity of their effects on properties.

3. The conditions of heat treatment, particularly solution temperature, and such structural factors as grain size might shift optimum Mo contents.

4. Properties were evaluated for rupture in 100 hours at 1600°F. The influence of Mo could change with temperature, time period or the use of

limited deformations rather than rupture as a criterion of strength.

A review of the composition of commercial alloys indicates that the role of Mo must be subject to a number of variables of the type discussed. There are British alloys with no Mo which reportedly have equivalent properties to American alloys. American alloys generally contain up to 4-percent Mo or 10-percent Mo. Waspaloy and M252 alloys have very similar rupture test properties even though their Mo contents are 4 and 10 percent. In this case, the 10 Mo in M252 alloy apparently is justified by improved characteristics other than creep-rupture properties.

The single most important variable in controlling strength of alloys of the type considered is the Ti + Al content. In view of the number of combinations possible as well as the influence of variations in other major alloying elements and the important effects of such trace elements as B and Zr, its becomes difficult to establish the influence of a single element such as Mo. A considerably more extensive research program than was carried out would be necessary to check whether or not the Mo contents of established alloys are optimum.

It was previously mentioned that Mo was believed to mainly retard microcracking in the grain boundaries during creep. Such microcracking is most severe at the interface between $M_{23}C_6$ carbides and the matrix. Molybdenum apparently promotes the formation of M_6C type carbides. When this type of carbide forms, the tendency to microcrack is very slight early in tests. The B and Zr act in the same way. It is, therefore, possible that Mo levels in commercial alloys were established before the role of B and Zr was understood and controlled. Thus it is possible that optimum Mo contents would be fairly sensitive to B and Zr contents.

The increase in rupture strength and ductility between 0 and 2 percent of Mo seems to be more than would be expected from the effect of Mo on microcracking for the B and Zr level in the alloy. It thus appears that some other mechanism is involved. This may be the usual solid solution and carbide precipitate effect.

Mo may also have an effect on the γ' reaction. The γ' precipitate is the main source of strength in the alloy. It has been reported that Mo does not enter into γ'. However, the authors are not aware of experiments designed to check the effect of Mo on the amount, size and stability of γ' precipitate particles.

The larger amounts of Mo in the alloys studied introduced center cracking during hot rolling. It should again be recognized that such characteristics are also probably a function of the Ti + Al content. B and Zr are also involved. One of the most important effects of B is to reduce surface cracking during rolling. Too much B or B + Zr introduced center cracking. Evidently the alloy could not tolerate as much B + Zr as was added when Mo contents were 8 to 12 percent. The severity of the direct rolling of ingots should be recognized. The cracking tendency under production conditions would be expected to be less.

It is important to recognize that optimum Mo contents must take into account other characteristics of alloys than creep-rupture properties. It is known, for instance, that M252 alloy is less susceptible to damage from brief overheats to temperatures above 1600°F than the 4 percent Mo alloys. M252 is also reputed to have better ductility and to be immune to notch sensitivity.

TABLE I
Reported Chemical Analysis of Heats
(Weight Percent)

Heat No.	Type Analysis	Mo	Co	Cr	Ti	Al	Ni	Si	Mn	C	B	Fe	Zr
1207	Wet	0	17.69	18.95	2.95	3.12	Bal	0.19	0.15	0.10	0.0052	0.12	---
	X-ray	0	17.67	18.95	2.95	--	--	--	--	--	--	--	*
1200	Wet	--	---	---	--	--	--	--	--	--	--	--	---
	X-ray	2.06	18.15	18.76	2.94	--	--	--	--	--	--	--	*
1202	Wet	--	---	---	--	--	--	--	--	--	--	--	---
	X-ray	3.82	17.91	18.88	2.94	--	--	--	--	--	--	--	*
1203	Wet	5.89	17.38	18.80	2.98	3.05	Bal	0.17	0.13	0.10	0.0052	0.12	---
	X-ray	5.89	17.39	18.79	2.98	--	--	--	--	--	--	--	*
1204	Wet	--	---	---	--	--	--	--	--	--	--	--	---
	X-ray	7.93	16.81	18.75	2.94	--	--	--	--	--	--	--	*
1205	Wet	--	---	---	--	--	--	--	--	--	--	--	---
	X-ray	9.85	16.43	18.79	2.96	--	--	--	--	--	--	--	*
1206	Wet	11.75	15.92	18.75	2.97	2.97	Bal	0.15	0.14	0.13	0.0056	0.12	---
	X-ray	11.75	15.94	18.73	2.96	--	--	--	--	--	--	--	*

* Radiation intensity same for all samples.

TABLE II

Creep-Rupture Data at 1600°F

Heat	Molybdenum(%) Nominal	Molybdenum(%) Reported	Stress (psi)	Rupture Time (hours)	Elongation (percent)	Reduction of Area (%)	100-hr Rupture Strength (psi)
1207	0	0	38,000	12.0	4	6	27,000
			28,000	77.2	2	5	
1200	2	2.06	38,000	67.5	14	16	34,500
			33,000	121.7	15	19	
1202	4	3.82	38,000	36.5	13	12	31,500
			32,000	91.5	11	14	
1203	6	5.89	33,000	57.9	5	5	31,000
			29,000	187.8	5	6	
1204	8	7.93	33,000	51.8	6	16	27,500
			29,000	83.7	10	19	
1205	10	9.85	33,000	14.9	26	65	23,500
			25,000	69.1	23	65	
1206	12	11.75	30,000	16.5	28	69	22,500
			24,000	67.2	27	63	

TABLE III

Results of X-Ray Diffraction Study of Alloy with 10-percent Mo

(Analysis of Residue Extracted from Heat 1205 with Hot Aqua Regia)

Measured Values		Identification of Measured Lines*			
d(h,k,l)	Intensity	σ	Cr_2O_3	Al_2O_3	TiC
3.62	vw		X		
3.48	vw			X	
2.66	w		X		
2.53	vw	X		X	
2.50	w				X
2.47	w	X	X		
2.42	w	X			
2.31	mw	X			
2.16	s	X	X		X
2.10	m	X			
2.07	vvw	X		X	
2.05	m	X			
2.00	ms	X			
1.96	s	X			
1.92	m	X			
1.86	vw	X			
1.81	vw		X		
1.78	vw	X			

* Check (X) shows that the measured line corresponds to data from the literature for the indicated phase.

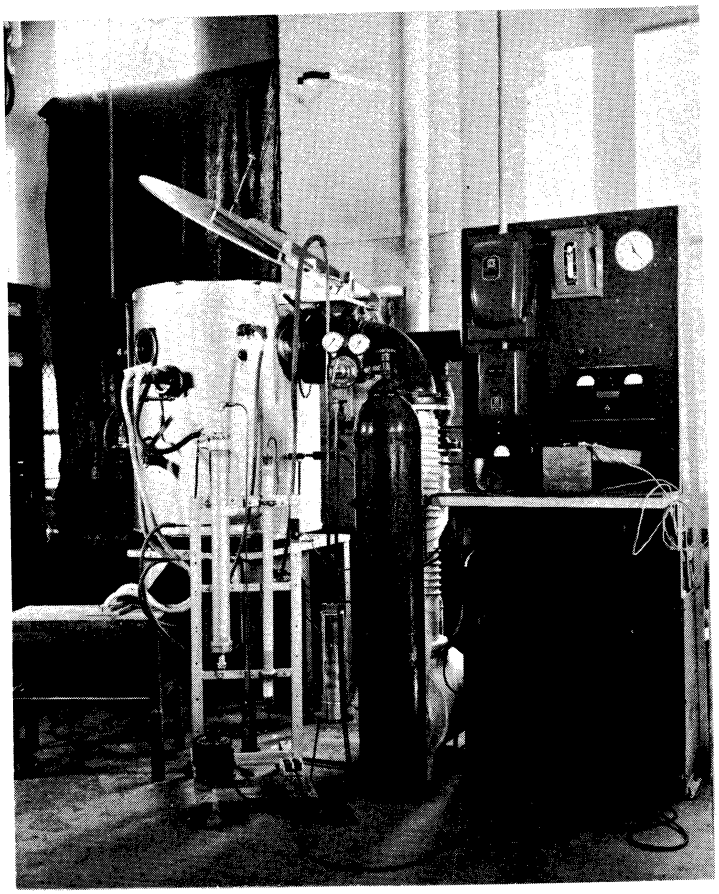
References follow:

σ -(Cr, Fe, Mo, Ni) - Allter, A. G., Journal of Metals, 6, 904 (1954)

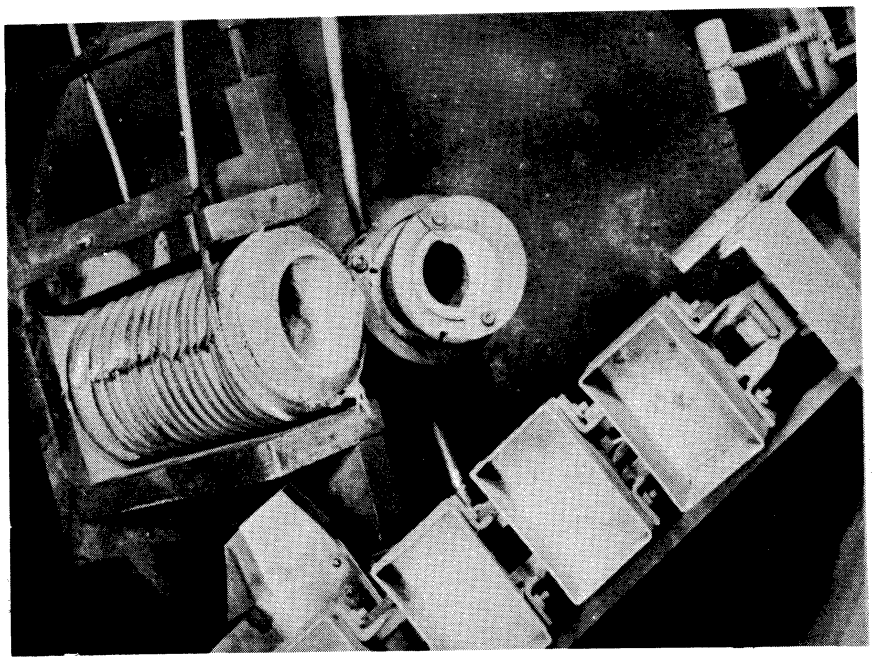
Cr_2O_3 - Swanson, Gilfrich, and Ugrinic, NBS Circ 539 Vol V, p. 22

Al_2O_3 - Swanson and Fuyat, NBS Circ 539 Vol II, p. 22

TiC - Rosenbaum, B. M., "X-Ray Diffraction Investigation of Minor Phases of 20-High Temperature Alloys", NACA TN 1580, 1948.



a. External view



b. Internal view of crucible, charge buckets and ingot mold.

Figure 1. - University of Michigan vacuum-melting furnace.

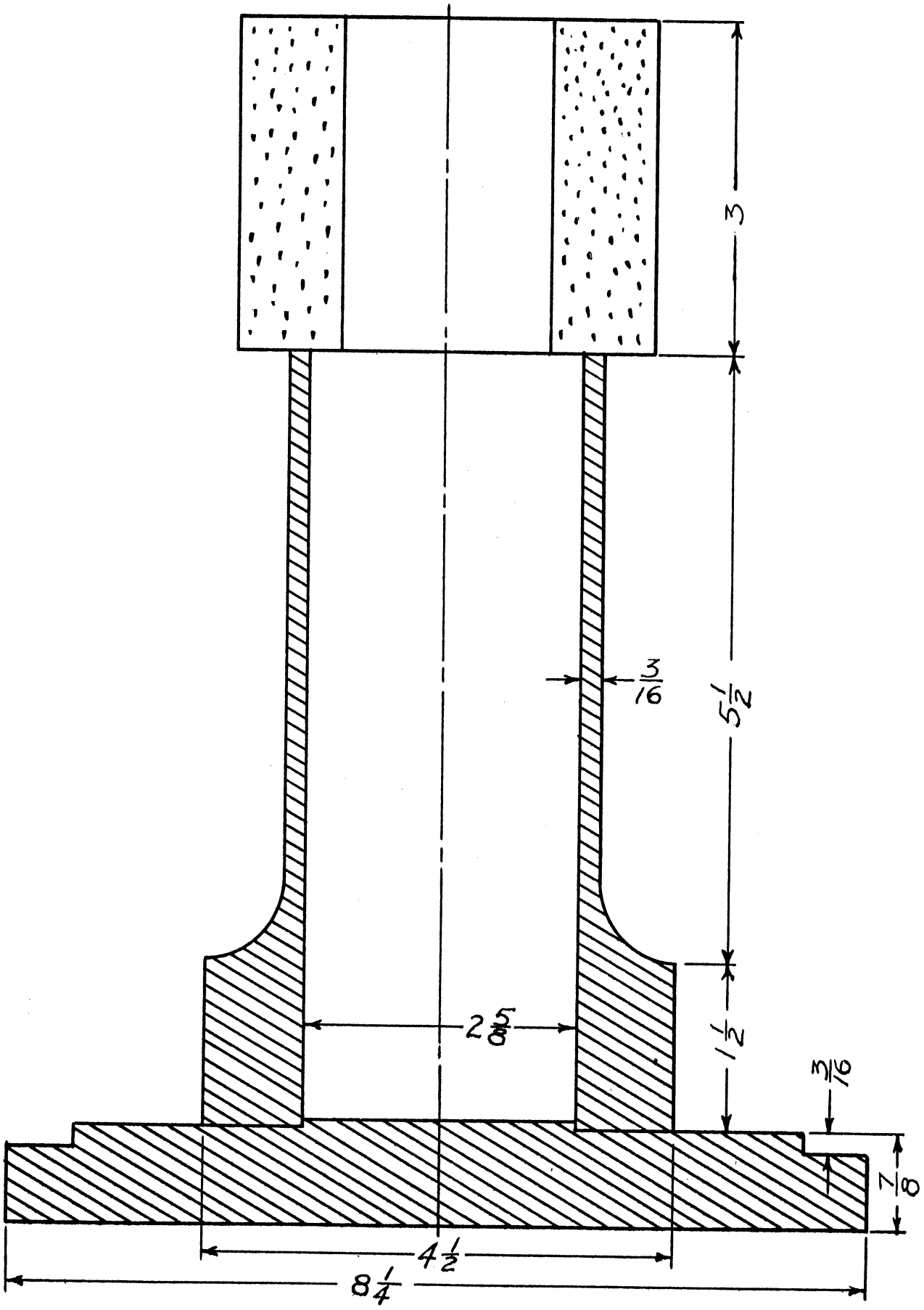


Figure 2. - Drawing of copper ingot mold and hot top arrangement for casting approximately 11 pounds of nickel-base alloy.

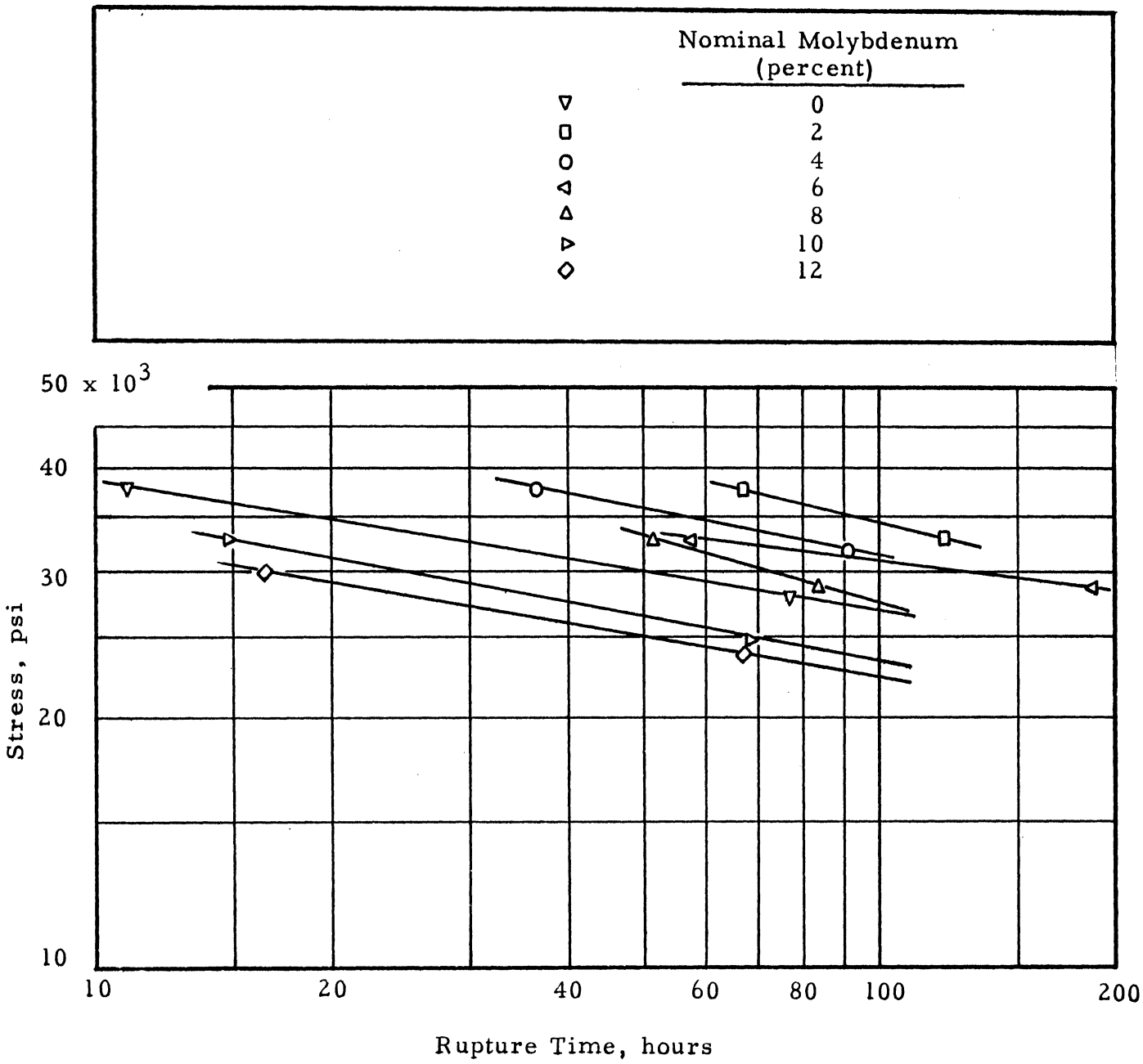
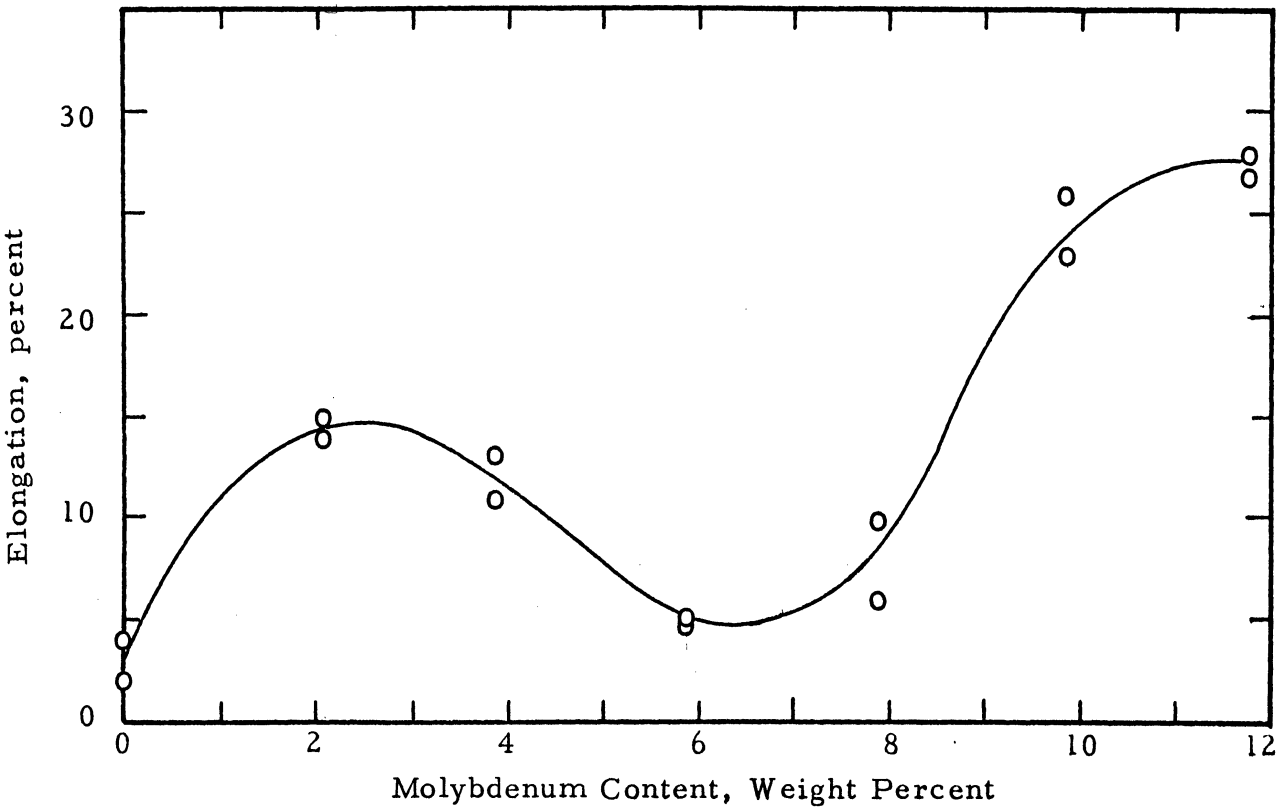
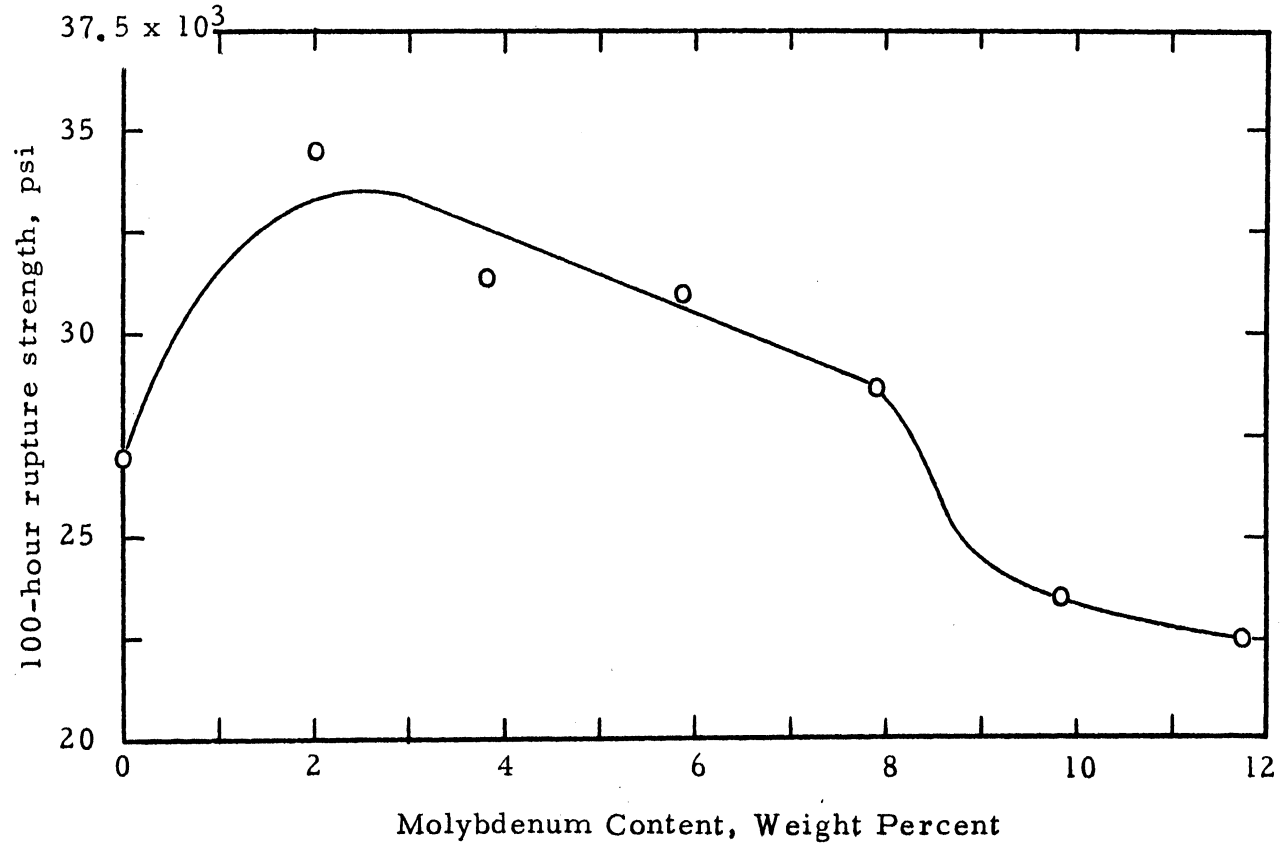


Figure 3. - Stress-rupture time curves for Udimet 500 alloy at 1600°F indicating the effect of variations in Mo content from 0 to 12 percent.



a) Elongation in rupture tests.



b) 100-hour rupture strength.

Figure 4. - Effect of Molbydenum content on the elongation in rupture tests and on the estimated 100-hour rupture strength at 1600°F for Udimet 500 alloy.

a) Solution treated 4 hours 2150°F AC plus 4 hours 1975°F AC

b) 4 hours at 2150°F, Air cooled + 4 hours at 1975°F, Air cooled + Aged 24 hours at 1550°F, Air cooled

c) 4 hours at 2150°F, Air cooled + 4 hours at 1975°F, Air cooled + Aged 24 hours at 1550°F + Aged 16 hours at 1400°F

Figure 6. - Effect of heat treatment on the microstructure of 12-percent Mo alloy (Heat 1206) X 11,000.

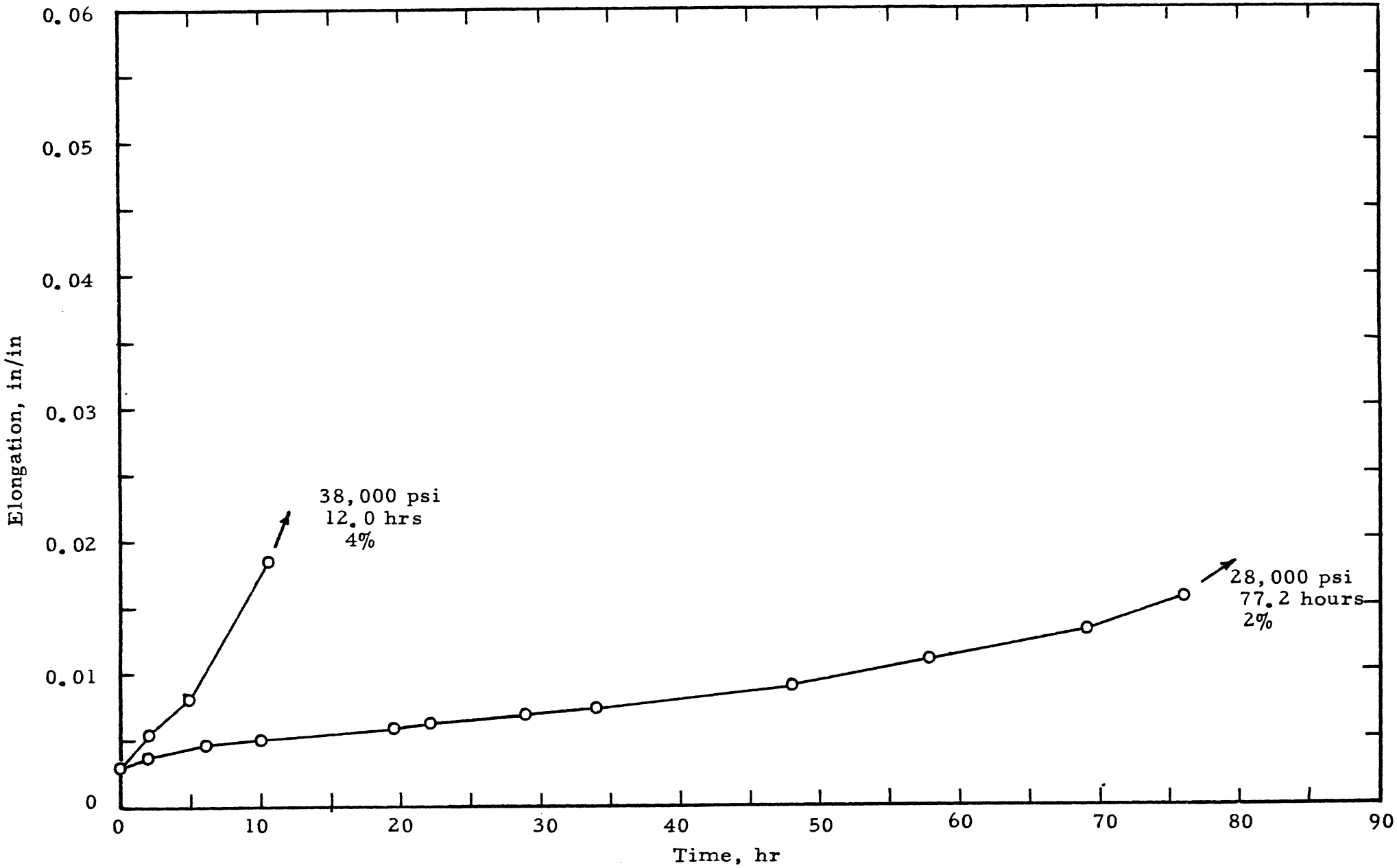


Figure 7. - Time elongation curves at 1600°F for alloy with 0 percent of Mo (Heat 1207).

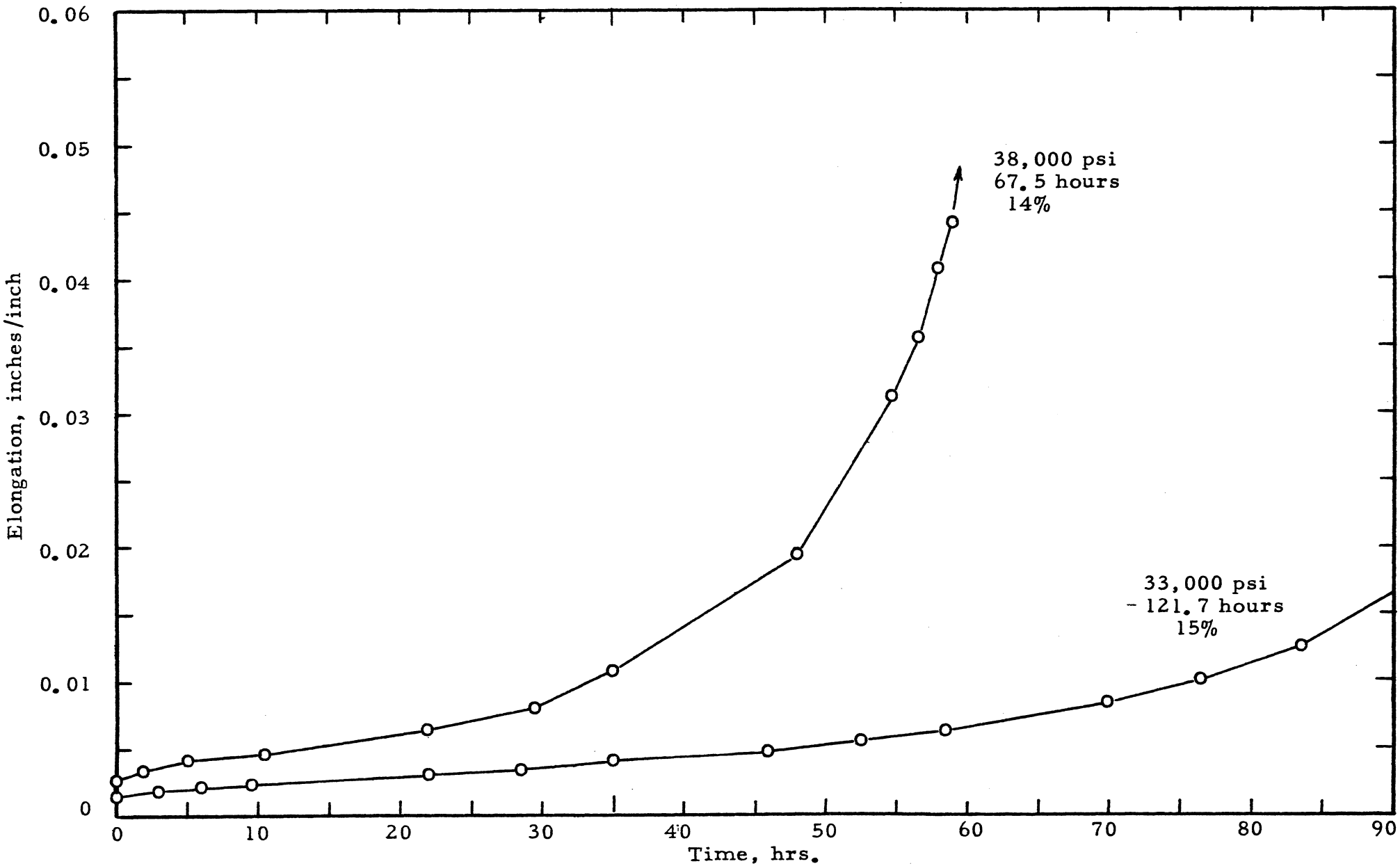


Figure 8. - Time elongation curves at 1600°F for alloy with 2 percent of Mo (Heat 1200).

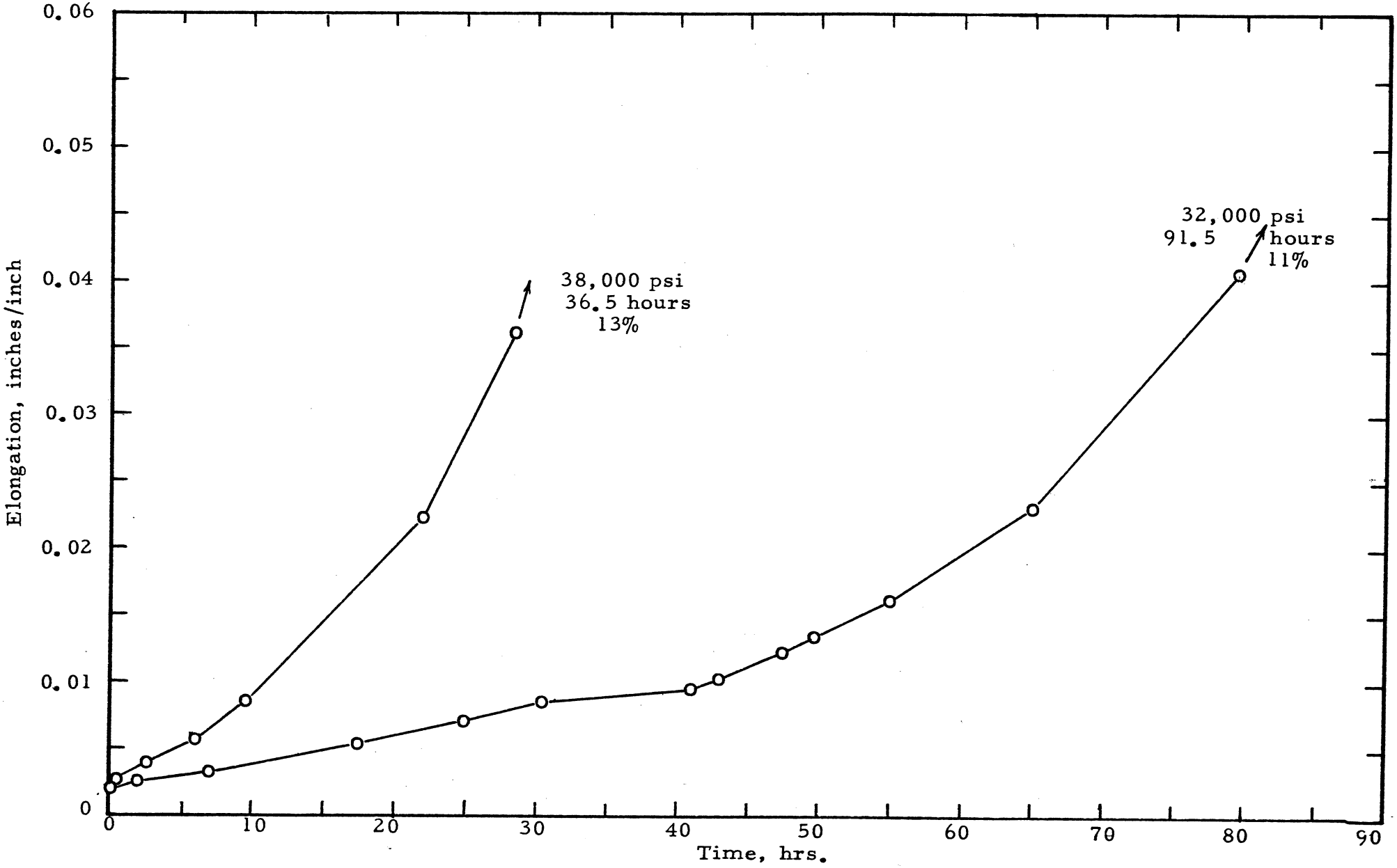


Figure 9. - Time elongation curves at 1600°F for alloy with 4 percent of Mo (Heat 1202).

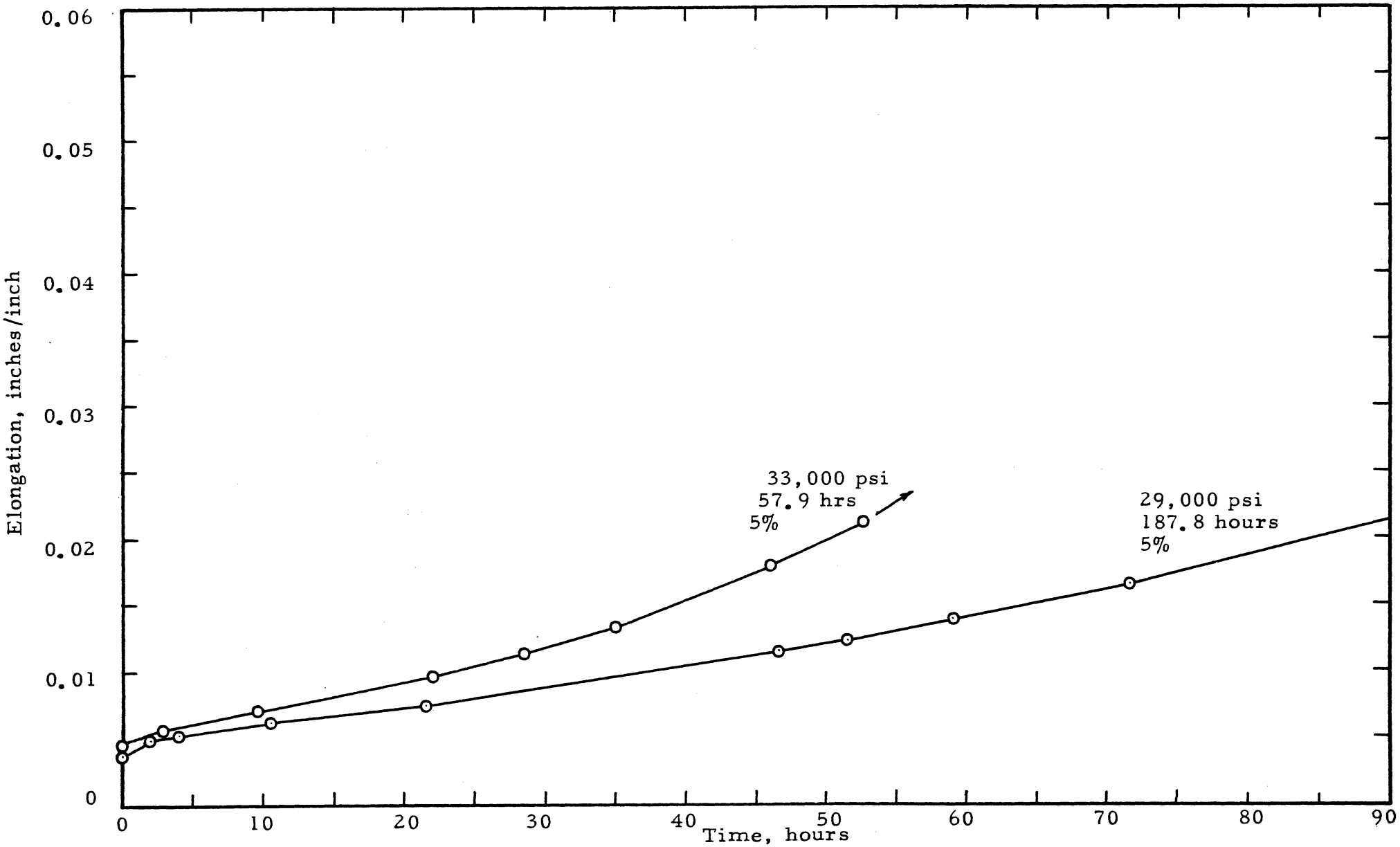


Figure 10. - Time elongation curves at 1600°F for alloy with 6 percent of Mo (Heat 1203).

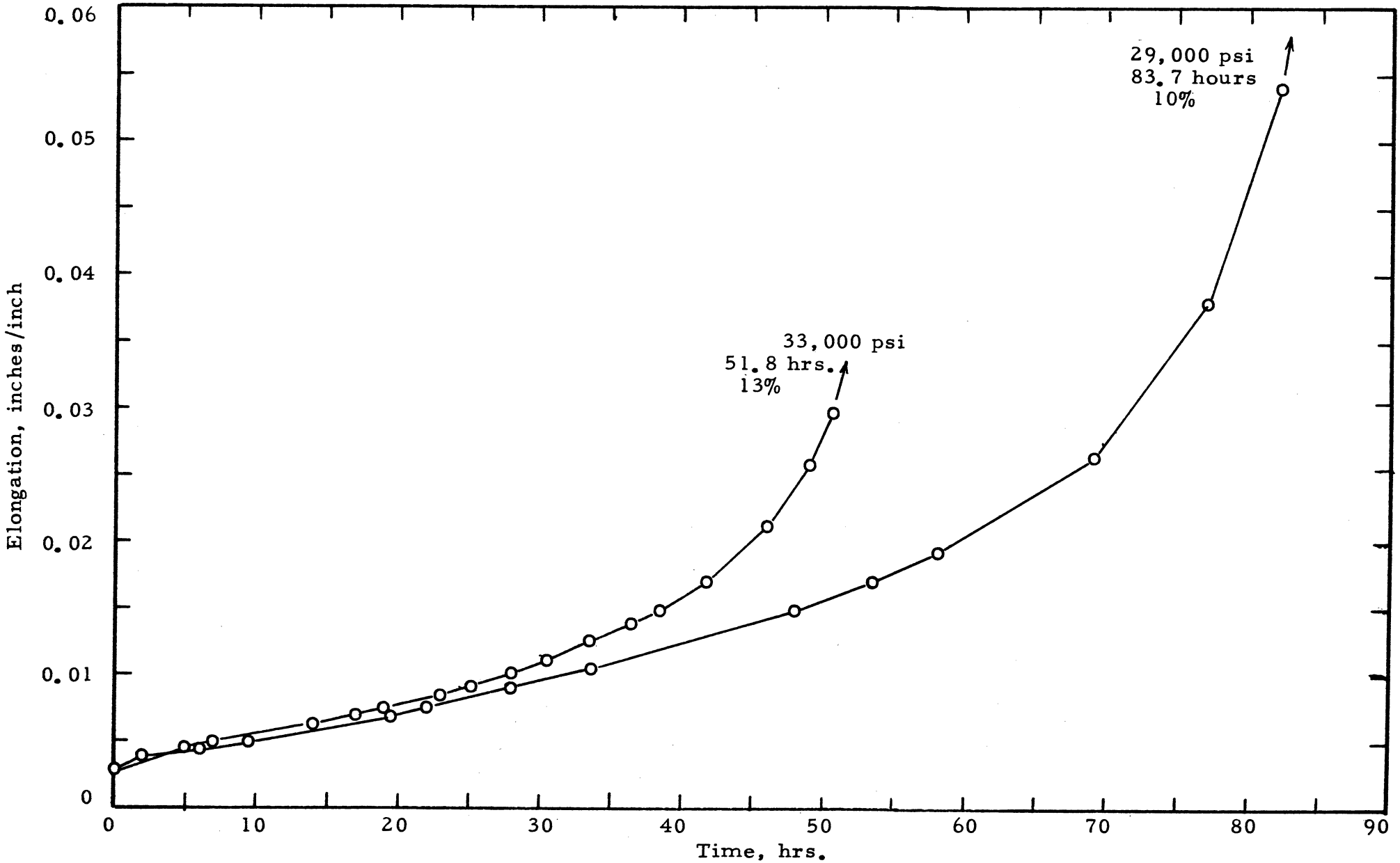


Figure 11. - Time elongation curves at 1600°F for alloy with 8 percent of Mo (Heat 1204).

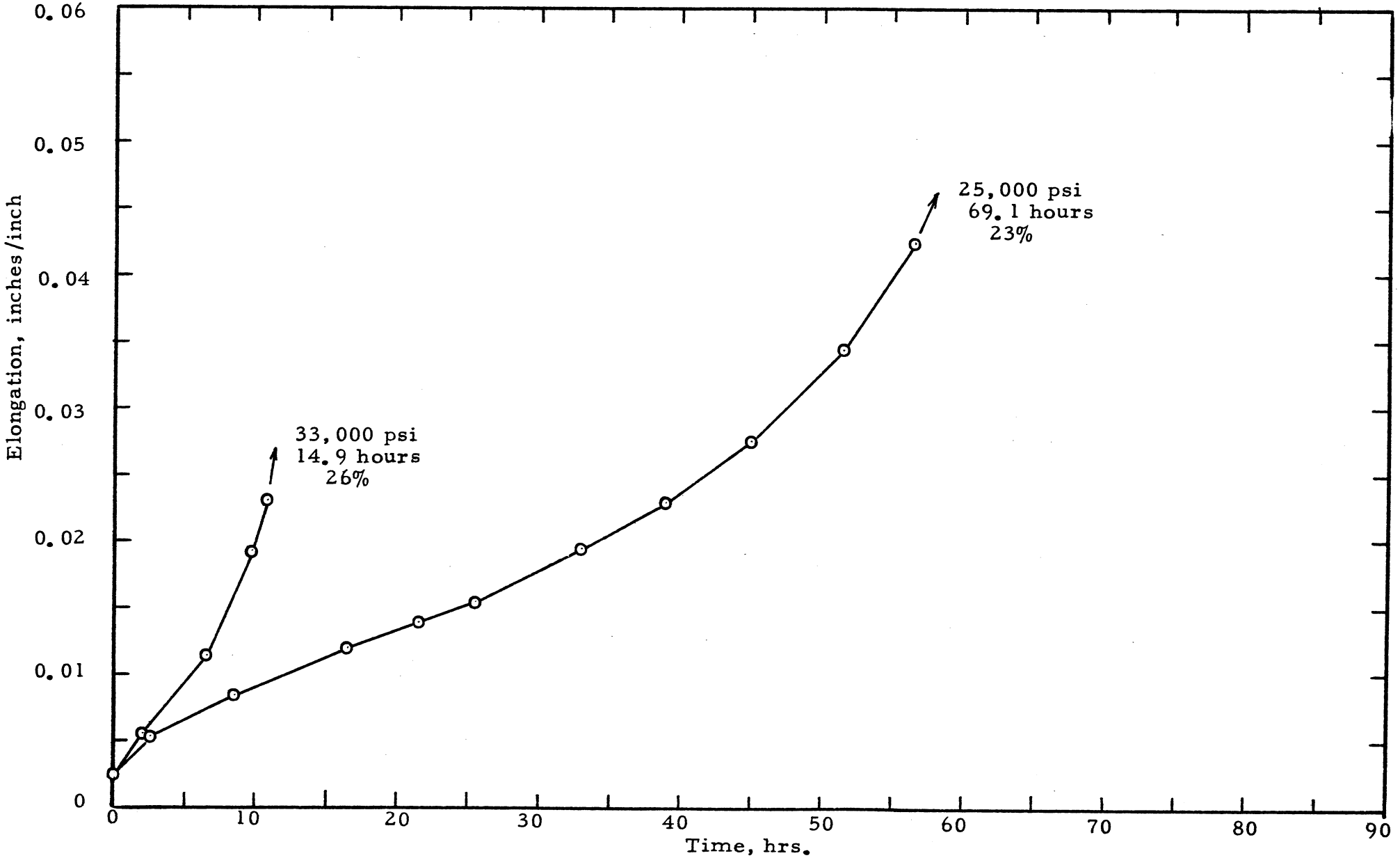


Figure 12. - Time elongation curves at 1600°F for alloy with 10 percent of Mo (Heat 1205).

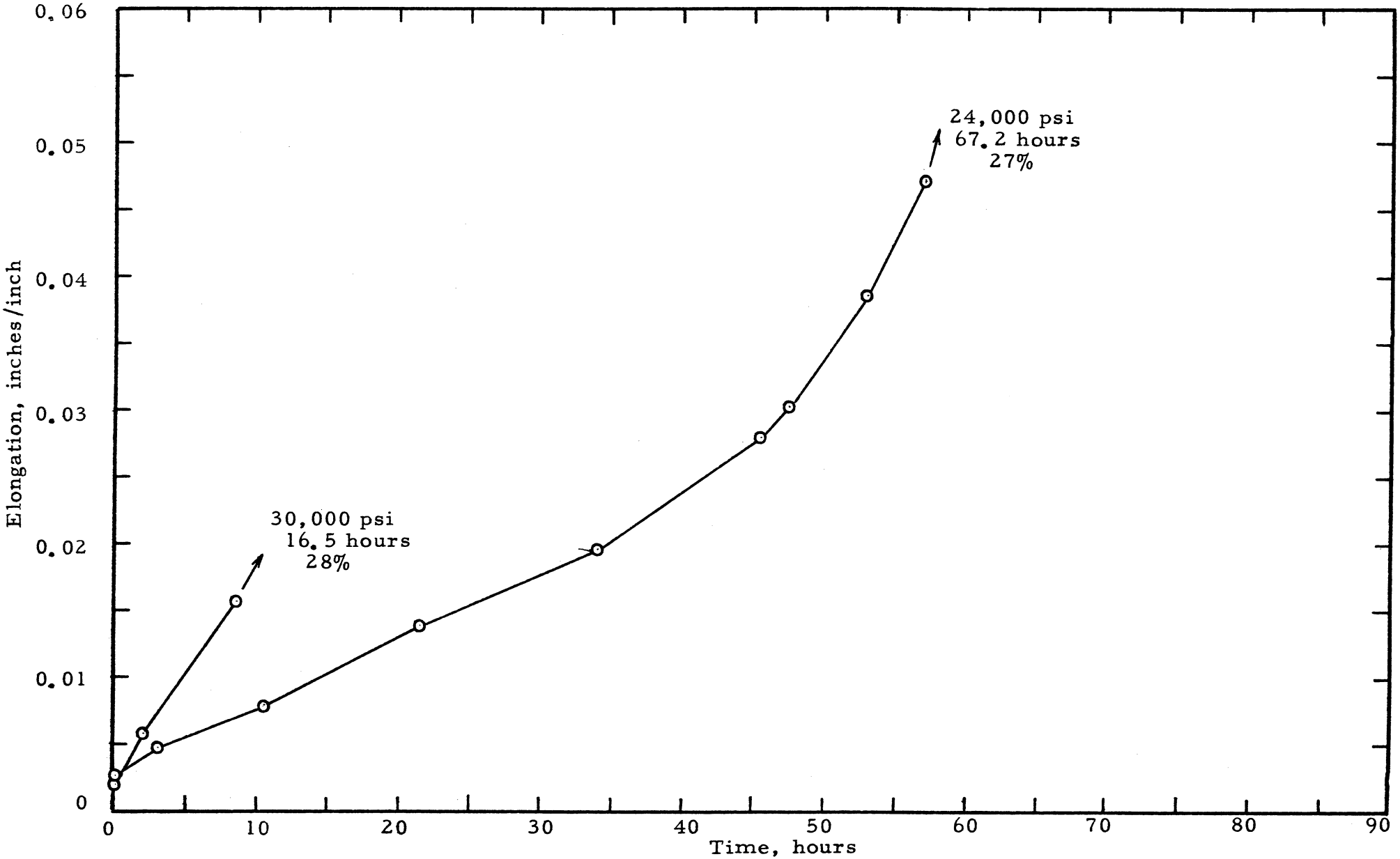


Figure 13. - Time elongation curves at 1600°F for alloy with 12 percent of Mo (Heat 1206).