Creep in Powder Metallurgy Products *

F. V. Lenel **

Introduction

Recent work has clearly shown that a fundamental understanding of the plastic properties of metals and alloys is possible only on the basis of dislocation theory. The plastic properties include the onset of plastic deformation, what is generally called the elastic limit, the increase in the stress necessary to cause plastic deformation with increasing strain, i. e. strain-hardening, the changes in the dislocation structure with temperature or recovery, and the plastic deformation of metals and alloys at a constant stress, usually observed at elevated temperatures and generally called creep. Those mechanical properties which are related to fracture, i. e. fracture strength and tensile strength, elongation and reduction in area and stress to rupture require for their understanding a knowledge not only of the formation and movement of dislocations, but also of the nucleation and propagation of cracks.

It must be admitted that even the qualitative explanation of certain phenomena in plastic deformation, e. g. strain hardening, is controversial. A quantitative calculation of plastic deformation parameters on the basis of dislocation theory has been possible only in a few cases. Nevertheless, since we are now quite certain that there is no alternative to the use of dislocation theory in the explanation of plastic deformation, we might as well take the plunge and try to grasp some of the phenomena involved at least in a qualitative way. In this lecture I will be concerned with the theory of creep and its applications to the steady state creep rate of certain powder metallurgy products. I have to make the theory as easy to understand as possible without oversimplifying matters too much.

Slip at Low Temperatures

Plastic deformation or slip in a metal lattice is not a uniform process, but takes place on particular crystallographic planes, the slip planes, in specific directions, the slip directions. These are usually the planes and directions of closest packing. Slip does not occur as a simultaneous motion with one part of the crystal moving along the slip plane with respect to the rest of the crystal. Instead it occurs as a consecutive process with slip starting at some point within the crystal and spreading out on the slip plane. Fig. 1 shows a section through a crystalline lattice with the slip plane A-A perpendicular to the plane of the figure. The left hand side of the crystal has undergone slip, the right hand side has not yet undergone slip. You see that the structure on the left and the right sides are identical. The consecutive nature of the slip process requires that between the slipped and the not yet slipped portion of the crystal there is a row of atoms which is shifting from one lattice position to another. This is seen in the center of the figure and is called a dislocation. Because the atoms in the line do not occupy normal lattice sites, they distort the surrounding crystal producing a strain-field. Plastic deformation should be described in terms of the number of dislocations and their rate of motion in the crystal.

In order to have appreciable plastic deformation one has to have numbers of dislocations

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** Rensselaer Polytechnic Institute, Troy, New York

moving extended distances in the lattice. This requires

1) generation of dislocations from discontinuities in the crystal lattice called dislocation sources. These may be segments of dislocations which are pinned at their ends, generally called Frank-Reed sources or grain-boundaries or free surfaces.

2) movement of the dislocations in the crystal lattice

What happens when we apply a stress to the crystal? Under its influence dislocations are generated from the dislocation sources in the crystal. In single phase material this stress which causes generation of dislocation from sources is the stress associated with yielding. The dislocations then move thru the crystal lattice until they meet some obstacle to dislocation motion, e. g. a grainboundary, second phase particle or arrays of suitably oriented other dislocations. This is illustrated in Fig. 2 All dislocations generated from a dislocation source subsequent to the first one, pile up behind the lead dislocation. Eventually, at low temperatures, no more dislocations can be generated from the dislocation source at the given stress because the previously generated dislocations which are now piled up reduce the effective stress on the dislocation source. At this point plastic deformation stops. In order to continue deformation, the applied stress will have to be increased. The stress which causes a certain amount of deformation or strain is the flowstress at this strain.

The strain associated with this flow stress is determined by the equation :

$$\begin{aligned} \text{Total strain (in/in)} &= \text{Density of sources} \\ &\times \text{Number of Dislocations from each source} \\ &\times \text{Area swept out by each dislocation} \\ &\times \text{Burgers Vector} \end{aligned}$$

The Burgers Vector is the unit slip associated with the movement of a dislocation and is of the order of interatomic spacing.

Creep

This is the picture for low temperatures, i. e. temperatures at which little or no recovery of the lattice takes place. This would be room temperature for iron, copper or aluminum, but on the other hand, a temperature well below room temperature for zinc or lead. At high temperatures a new phenomenon which we call recovery must be taken into consideration. At these temperatures, the dislocations in the piled up group are no longer anchored in their slip plane but may leave it. The most important recovery process contributing to creep is dislocation climb which is illustrated in Fig. 3 This figure is very similar to figure 1, except that I have added another type of defect in the crystal lattice, i. e. unoccupied lattice sites, called lattice vacancies. A few vacancies exist even at low temperatures, but they are not important for plastic deformation, because there are so few of them and because they cannot readily move from one lattice position to another. At higher temperatures both the number of lattice vacancies and their mobility increase exponentially. The movement of these lattice vacancies is what causes diffusion and, as you know, diffusion becomes appreciable only at high temperatures. In the figure I have indicated a vacancy which has moved right next to the slip plane. If now the vacancy moves into the position of the last atom in the extra row of atoms of the dislocation and the same thing happens to all the other last atoms in the dislocation line, the slip plane of the dislocation moves up one step. This we call dislocation climb; the dislocation has moved not in its normal manner along the slip plane, but in a direction normal to the slip plane. Moving only one step will not get the dislocation unpinned from its pile-up, but if this climb takes place several times, the dislocation eventually becomes free of the pile-up.

This process is further illustrated in Fig. 4 which shows a number of dislocation sources generating dislocations on different slip planes.¹⁾ The arrows indicate how the dislocations climb out of their slip plane. Eventually they may interact with a dislocation coming from another slip plane and, having opposite sign, that is opposite structure, become annihilated.

The important fact as far as further slip is concerned is that the dislocations leave the pile-up. As soon as this happens, the dislocation source will be able to generate other dislocations and plastic deformation or slip continues even at a constant stress. What we are particularly interested in is the rate at which further flow takes place, i. e. the strain rate. We can calculate the rate at which such recovery processes as dislocation climb take place on the basis of diffusion considerations. This rate of recovery in turn governs the rate at which new dislocations can be generated. We can then write an equation for the strain rate :

$$\begin{aligned} \text{Strain rate (in/in/unit time or \% /unit time)} = & \\ & \text{Number of sources per unit volume} \\ & \times \text{rate of generation of dislocations from each source} \\ & \times \text{area swept out by each dislocation} \\ & \times \text{Burgers vector of dislocation} \end{aligned}$$

In creep experiments this strain rate is the steady state creep rate.

We must now consider another type of creep in which the creep rate is not dependent upon the rate of recovery which permits generation of new dislocations from a fixed number of sources. In materials with a very large area of grainboundary per unit volume, in other words in very fine-grained materials the grainboundaries may become the most important source of the dislocations, causing creep. Any part of the large grainboundary area may act as a dislocation source. Dislocations leaving a given small area of grainboundary which acts as source pile up within the crystal and prevent generation of further dislocations from this small area. It may now not be necessary to wait until dislocations leave this pile up and further dislocations can be generated from the small area. Instead another area of grainboundary may take over as a dislocation source and start generating dislocations. In this type of creep the creep rate is not governed by the rate of recovery but is determined by the rate of generation of dislocations from the total grainboundary area. This is the explanation based on dislocation theory of the creep phenomenon usually called grainboundary sliding.

Creep in Dispersion Strengthened Materials Governed by Rates of Recovery

For alloy systems in which creep rates are governed by rates of recovery, creep may be substantially slowed by slowing the recovery process. One way of doing this is by dispersing a second phase of suitable size, shape and distribution of the particles in the matrix metal or alloy. Such a dispersion will increase the distance over which dislocation climb occurs and therefore slow up recovery²⁾. How second phase particles interact with the movement of dislocation is illustrated in Fig. 5 and 6.

Figure 5 applies to creep at low stresses. The forked line on the left represents a dislocation source which is able to generate dislocations under the influence of an applied stress. The slip plane in which the generated dislocations move is the plane of paper. The stress is of such magnitude that the second phase particles act as barriers against which the dislocations are piled up. Practically no plastic deformation can take place at low temperatures. At elevated temperatures, however, the dislocations can climb out of the slip plane. This is shown in B where the slip plane is now horizontal and perpendicular to the plane of the paper. The letters FR stand for Frank-Read and represent sources which generate dislocation. In order to make it possible for sources to generate dislocations and cause creep, the dislocations must

Fig. 1 Section through crystalline lattice with dislocation on slip plane A-A.

Fig. 2 Dislocation source generating dislocations, which pile up against barrier.

Fig. 3 Dislocation climb by movement of vacancy.

Fig. 4 Dislocations generated by dislocation sources on different slip planes climbing out of their slip planes and being annihilated by interaction with other dislocations.

Fig. 5 Model showing interference of dispersed particles with creep at low stress; a) Dislocation source with slip plane in plane of paper; b) dislocations generated under the influence of stress from sources (Frank-Read) slip plane horizontal and perpendicular to plane of paper; c) view of dislocation line from the direction of motion.

Fig. 6 Model showing interference of dispersed particles with creep at high stresses; a) dislocations looping around particles and dislocation loops pinching off, slip plane in plane of paper; b) climb of dislocations around particle, slip plane horizontal and perpendicular to plane of paper.

climb out of the slip plane and climb over the second phase particles of height h . This height h of the particles determines the creep rate at a given temperature and stress.

The situation at higher stresses is illustrated in Fig. 6. On the left hand side, the plane of the paper is again the slip plane and the line represents a dislocation, generated by a source and moving against the second phase particles. In this case the stress is sufficient so that the dislocation line bows between the particles and eventually forms a dislocation loop around the particle while the dislocation line moves on. Even in this case, the amount of plastic deformation at room temperature is very limited because the successive loops formed around the particles rapidly inhibit generation of further dislocations from the source. In order to get creep at elevated temperatures it is again necessary for dislocation climb to occur. In this case it is dislocation loops looped around the particles which must climb. This is illustrated on the right hand side of the figure, in which the slip plane is again horizontal and perpendicular to the plane of the paper. We see 2 symbols of dislocations which represent loops around the second phase particle. The third dislocation loop is climbing. The distance a dislocation loop must climb before the next loop can be pinched off is determined by the spacing between second phase particles and the applied stress. These therefore control the rate of steady state creep.

Creep in Recrystallized SAP

The dispersed phase system in which a detailed investigation of creep has been made is the dispersion of aluminum oxide in aluminum, the so-called SAP-type alloys. These alloys are produced by compacting, hotpressing and extruding aluminum powder. The thin coating of oxide, approximately $100\text{\AA}$ thick on the surface of the aluminum particles is broken up during processing and forms the dispersion of second phase aluminum oxide particles in the aluminum matrix. The spacing of the oxide particles can be controlled by the size of the aluminum particles which are used to produce the extruded compacts. SAP is a particularly good material for investigating the creep characteristics of dispersion strengthened materials, because the size and the spacing of the oxide particles is extremely stable even up to the melting point of aluminum. In order to compare calculated and measured creep rates it is necessary to use SAP type alloys with a relatively large grain size for reasons which will be discussed later. SAP type alloys with large grain sizes can be produced by severely cold-working and then recrystallizing the original extrusions. This is possible with those alloys which have spacings between particles coarser than 0.5 microns. The very strongest SAP type alloys with the finest spacings cannot be made coarse grained by recrystallizing.

The results of creep tests on two of these recrystallized SAP-type alloys, called AT-400 and RP 15-30⁹⁾ are shown in Fig. 7. The plot shows the logarithm of the steady state creep rate per minute as a function of the logarithm of the stress in pounds per square inch. The creep rates of the two SAP type alloys are compared with the creep of high purity aluminum at 500°C . We see that the creep rate of the SAP type alloys is many orders of magnitude slower than that of pure aluminum. The stress range is such that the steady state creep rate should be calculated using the high stress model. The calculation permits us to predict the dependence of creep rate upon stress, upon spacing and upon temperature. The creep rate should be proportional to the fourth power of stress and this is found in the experiment as the figure shows. The creep rate should be proportional to the square of the spacing between particles. The spacing is approximately 1 micron in the AT-400 alloy and approximately 6 microns in RP 15-30 alloy. The ratio of the creep rates is 32 in good agreement with the theory. The temperature dependence of the creep rate is not shown in the figure.

However, this temperature dependence is found to be about the same as the temperature dependence of diffusion in aluminum which is what one would expect if creep is controlled by dislocation climb.

If one wants to calculate the absolute values of the steady state creep rate one must know the number of dislocation generating sources per unit volume. Estimating the number of these sources in accordance with the number usually found in annealed metals, gives calculated creep rates much higher than those found experimentally. Why there are so few dislocation generating sources in these SAP alloys is now being investigated at R. P. I. on the basis of transmission electron micrography of very thin sections.

Creep in Dispersion Strengthened Lead

As was pointed out before, recovery takes place in lead and zinc at room temperature. These two metals and many of their alloys have, therefore, appreciable creep rates at room temperature. The design of lead alloys which are creep resistant at room temperature is of great practical importance. The Lead Industries Federation is sponsoring a project at Rensselaer Polytechnic Institute to develop such alloys by powder metallurgy. As in the case of aluminum alloys it should be possible to reduce the steady state creep rate of lead alloys by dispersing in them a finely distributed second phase. The method used at R.P.I. is analogous to that suggested by Towner⁴⁾ for aluminum alloys, in which the dispersed second phase is a metal or intermetallic compound. A suitable system for lead as the matrix metal is the lead-copper system. Copper is practically insoluble in solid lead, but has increasing solubility with increasing temperature in liquid lead. When a lead-copper alloy is heated to a temperature where the copper is completely dissolved in the liquid lead and atomized from this temperature, a powder results in which each particle consists of a very fine dispersion of copper in a lead matrix. Most of the work has been done on an atomized alloy powder with 5% Copper. The powder is compacted and extruded at temperatures between 100-225°C. The microstructure of a lead-copper powder extrusion is shown in Fig. 8. The copper particles in the extrusion are practically spherical, they have an average diameter of 0.8 microns and an average interparticle spacing of 1.6 microns. Some of the results of creep tests on these alloys are shown in Table 1. They may be compared with creep tests on conventional lead alloys which are shown in Table 2. No attempt has yet been made to calculate steady state creep rates of these alloys on the basis of dislocation theory, but work in this direction is planned. There seems to be little doubt that the improved creep strength of the copper-lead alloy must be attributed to the effect of the fine dispersion of copper particles in the lead matrix.

Table 1 Steady State Creep Rate of 95% Lead-5% Copper Powder Extrusions at 25°C.

Extrusion Temperature °C	Creep Stress Psi	Steady State Creep Rate %/Year
120	2600	27.6
120	3150	44.8
120	4050	184.0
230	3150	35.0
230	4050	113.0

Table 2 Comparison of Steady State Creep Rates of Rolled Lead Alloys and of a 95% Lead-5% Copper Powder Extrusion.

Alloy	Temperature °C	Creep Stress Psi	Steady State Creep Rate %/Year
6% Antimonial	30	860	100
Chemical Lead	30	700	100
Common Lead	30	730	100
95% Pb - 5% Cu Powder Extrusion	25	4050	113

Creep in Fine-grained SAP

In the theoretical section we discussed a type of creep prevalent in very fine grained materials, in which it is not the rate of recovery, but the rate of generation of dislocations from the grain boundaries which controls the rate of steady state creep. Curiously enough this type of creep, often called grainboundary sliding, is found in SAP-type extrusions, as extruded. In this condition the alloys have elongated grains whose thickness is of the order of five microns. Creep rate studies were made by Ansell and Weertman²³ on one of these alloys, MD 2100, having an average spacing between particles of 0.5 microns. The spacing is therefore about one half that of the AT 400 alloy, whose creep behavior after recrystallization was shown in Fig. 7. Careful examination of replicas under the electron microscope has shown that the spacing is not influenced by the cold working and annealing operations necessary to cause recrystallization. Creep rates of this alloy in the as extruded and in the recrystallized condition are compared in Table 3. It is seen that the recrystallized

Table 3 Comparison of Steady State Creep Rates of MD 2100 Alloy in the Extruded and the Recrystallized Condition at 500°C.

As Extruded		Recrystallized	
Stress psi	Creep Rate in/in/min	Stress psi	Creep Rate in/in/min
3,000	6×10^{-5}	3,000	$< 10^{-8}$
4,100	1.2×10^{-4}	4,100	$< 10^{-8}$

alloy does not creep at all within the limit of experimental accuracy even at 500°C, i.e. 160° under the melting point of aluminum and stresses as high as 4000 psi.

In the as-extruded, finegrained condition, on the other hand, the steady state creep rate at 500°C increases with increasing stress and reaches 1.2×10^{-4} per minute or 17% per day at a stress of 4100 psi. This creep must be attributed to the fine grain size of the alloy favoring a creep process in which the creep rate is controlled by the generation of dislocations from grainboundaries. Even though the creep is several orders of magnitude faster than for recrystallized material, it is still very much slower than for pure aluminum and most aluminum alloys.

Creep in Zinc-Powder Extrusions

Another example of a fine-grained metal powder extrusion, in which creep is controlled by the rate of generation of dislocations from grainboundaries, is that of zinc-powder extrusions. Creep by grainboundary sliding is particularly characteristic for close-packed hexagonal zinc. The properties of extrusions made from atomized zinc powder were investigated in connection with a study sponsored by the American Zinc Institute at Rensselaer Polytechnic Institute. When an atomized zinc powder with an average powder particle size of 50 microns is cold compacted and extruded at 300°C, the resultant wire has a very fine grainstructure with an average grainsize of only about 5 microns. Extrusions from cast zinc ingots have a strong tendency to develop a very coarse grain size when they are either extruded at high temperatures or annealed at these temperatures after extrusion. In contrast the grainsize of the powder extrusions is quite stable. Because of their stable grainsize the room temperature strength and ductility of the powder extrusions is also relatively high and independent of extrusion or annealing temperature, while the strength and ductility of cast zinc extrusions decreases rapidly with increasing extrusion or annealing temperatures.

A comparison of the creep rate at 25°C and 8000 psi stress of a zinc powder extrusion with that of a series of extrusions from cast billets with various grainsizes is shown in Fig. 9. In accordance with the custom of zinc metallurgists the reciprocal or inverse creep rate

Fig. 7 The logarithm of the steady-state creep rates of AT 400 and RP 15-30 recrystallized Al-Al₂O₃ SAP-type alloys at 500°C and pure aluminum at 433°C plotted versus the logarithm of the applied stress.

Fig. 8 Microstructure of 95% Pb-5% Cu powder extrusion. 1000X

Fig. 9 Inverse steady state creep rate at 8000 psi stress in days/% of extrusions from atomized zinc powder and from cast billets plotted vs average grain diameter of extrusions.

in days per % instead of the usual creep rate in % per day is plotted vs. the grain size of the extrusions. It is evident that the powder extrusion creeps much faster (0.05 days/%) than the extrusions from cast ingots and that the creep rate is clearly grain size controlled.

It must be assumed that the particles of atomized zinc powder just like those of atomized aluminum powder have an oxide coating on their surfaces and that this coating is broken up during extrusion into particles of zinc oxide in a zinc matrix. However, the creep rate of these dispersions of zinc oxide in zinc seems to be controlled entirely by zinc grain boundary creep. In contrast to the aluminum case the dispersed zinc oxide particles do not seem to affect the creep rate.

Creep in Zinc Alloy Powder Extrusions

The study of zinc powder metallurgy like that of lead powder metallurgy is directed towards the development of zinc alloys which are creep resistant at room and slightly higher temperatures. The very fast creep of the zinc powder extrusions is an interesting example of creep due to the generation of dislocations at grain boundaries, but from a practical point of view, it is not what we looked for. Being at heart, much more a practical than a theoretical metallurgist I would hate to end this lecture on such a negative note. Fortunately, this is not necessary since I can report some preliminary results on the creep of zinc alloy powder extrusions, which show excellent creep resistance. These extrusions were produced from a powder containing approximately 1 % of Copper and 0.15 % of titanium. The alloy composition was developed several years ago at the laboratories of the New Jersey Zinc Company. Sheet rolled from cast ingots of this alloy has much better creep resistance than most other

zinc alloys. It therefore seemed interesting to compare the creep resistance of extrusions made from powder of this composition with that of extrusions from cast billets. Some of our first results are shown in Table 4.

The data show that at 25°C and a stress of 16,000 psi, the extrusion from a cast billet has an inverse creep rate of 9.8 days/%, while the creep rate of the powder extrusion is too small to be measured. In order to get

Table 4 Inverse Creep Rates of Extrusions from 0.95% Cu, 0.13% Ti, bal. Zn Alloys at 25°C

Extruded from	Stress, psi	Inverse Creep Rate, days/%
Cast Billet	16,000	9.8
Powder Compact	16,000	> 1000
Powder Compact	24,000	114

measurable creep rate for the powder extrusion the stress must be increased to 24,000 psi. At this stress the inverse steady state creep rate of the powder extrusion is 114 days/%.

The reasons for the excellent creep resistance of the zinc-copper-titanium alloy powder extrusions are, unfortunately, not yet known. There are a number of possibilities, but in the absence of definite experimental evidence it seems futile to do much speculating. Work on these interesting extrusions is continuing and eventually an answer should be forthcoming. In spite of my inability to give you a suitable theoretical explanation for the creep rate in zinc alloy powder extrusions I hope I have been able to demonstrate that we have at least a beginning for the explanation of creep in powder metallurgy products by means of dislocation theory.

Acknowledgement

I should like to thank my colleague Dr. G. S. Ansell who assisted me in the preparation of this lecture.

References

- 1) Figure 4 from J. Weertman : Trans. AIME. 218 (1960), 207 ; see also J. Weertman, J. Appl. Phys., 26 (1955), 1213 and 28 (1957), 362
- 2) G. S. Ansell and J. Weertman : Trans. AIME. 215 (1959), 838
- 3) F. V. Lenel and G. S. Ansell : A Theory of Dispersion Strengthening, in Powder Metallurgy, edited by W. Leszynski: New York, (1961), 267
- 4) R. J. Towner : Metal Progress, 73 No.5 (1958), 70,

著 者 紹 介

Dr. Fritz V. Lenel はアメリカ、ニューヨーク州の Troy 市にある Rensselaer Polytechnic Institute の Metallurgical Engineering Dept. の教授で、Powder Metallurgy, Diffusion その他の講義を担当するとともに、同大学に所属する Powder Metallurgy Laboratory の Head として、研究の計画、推進、指導等を行なっている。彼はドイツ人で、1931年に Heidelberg University で Ph. D をとり、その後1933年まで Goettingen University で研究にたずさわった後、アメリカにわたって1937年まで Charles Hardy, Inc. に、1947年まで Moraine Products Div., General Motors に勤めた。この間に粉末冶金の工業的利用についての豊富な経験を経て、1947年現在の R.P.I. に移った。1955年には ASTM の総会で "Powder Metallurgy-Now" という演題のもとに Gil-

lette Memorial Lecture を行ない、また ASTM の Committee B-9 (粉末冶金製品関係) や AIME の Powder Metallurgy Committee の Secretary や Chairman をつとめて、粉末冶金の学界、業界において相当活発な活動を行なっている。1960年の第一回国際粉末冶金会議においても、Dr. C. G. Goetzel, Dr. H. H. Hausner とともに Co-chairman としてその計画遂行に大いに寄与している。

現在彼の P. M. Labo. で強力に進められている研究テーマは大別して次の2つである。

1) Mechanism of Strengthening in Dispersion

Strengthened Alloys

2) Mechanism of Sintering of a Single Metal Compact.

特に1)については、SAP の yield strength, creep strength に対する理論的解析と実験的考察とを骨子として研究が進められており、さらにこの応用として、Ni, Pb, Zn をベースとする分散型合金の諸特性が多角的に研究されている。ここに寄せられた論文は、これらに関するこれまでの研究結果を総括してまとめたものである。

(名古屋工業技術試験所 林 悦雄記)

本会事記

第二十回役員会議事録

日 時 昭和36年9月11日(月) 9時30分より

場 所 明治大学大学院

出席者 岩瀬慶三 外21名

議 事

1 P. R. 講演会について

- (1) P. R. 講演会を正式には粉末冶金製品講演会と呼ぶことに決定。
- (2) 講演プログラムの順序を決定、講演時間25分、質疑時間5分とすることに決定。
- (3) 各工業会は P. R. 用に各工業会員の名簿、製品名等を印刷したものを講演要旨の後に綴じ込むことに決定。この事務は東京事務局で行ない、印刷物を20に京都に必着することとする。
- (4) 講演要旨は1500部印刷、京都本部100、会場200、各工業分計1200部(東京事務局宛)分配する。
- (5) 各工業会への分配は東京役員会に一任する。1200以上必要とする場合は、京都に20日までに連絡する。なお、概要、名簿等の印刷代その他は各工業会の負担とする。

2 秋 期 大 会

鹿取委員長より特別講演、一般講演、見学会、懇

親会について報告。会長より一般講演に粉末冶金プロパーの講演が少ないので、今後多くするよう要望があった。

3 名簿印刷について

本部より仮ズリが配布され、訂正を加えた。

4 次 回 分 科 会

12月4日、5日に分科会を開催する。

5 外 人 会 員 の 件

Association des Ingenieurs et Techniciens Mecanicens et Electriciens de Yougoslavie よりの申し込みを機会に、外人会員について一応次の如く決定。

	入会金	会 費
個 人	5 \$	10 \$
法 人	50 \$	200 \$

なお、この会費は協会会則にないので法的処置は今後検討することになった。

6 役 員 (顧問) 変更の件

顧問三浦倫義氏(株)日立製作所が小野健二氏に変更の旨当社より申出あり、同氏を承認した。

7 入 会 の 件

特別会員として日本電池(壱口)と正会員12名(7月~8月末)入会を事務局より報告、承認した。