1994

A metallographic study of 18th century woodworking tools from the Williamsburg collection

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TITLE:

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DATE: May 29, 1994

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by

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A Thesis

Presented to the Graduate and Research Committee

of Lehigh University

in Candidacy for the Degree of

Master of Science

in

Materials Science and Engineering

Lehigh University

May 18, 1994

This thesis is accepted and approved in partial fulfillment of the requirements for the Master of Science.

May 15, 1994

Date

Thesis Advisor

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Abstract

In order to provide some insight into questions concerning the identity and quality of woodworking tools, a group of these tools from the Colonial Williamsburg Museum was brought to Lehigh University for in depth metallurgical examination. The museum selected a total of 33 eighteenth century woodworking tools for our consideration. The tools were subjected to detailed metallographic analysis which emphasized light optical microscopy (LOM) of both the macrostructure and microstructure and to microhardness measurements as a function of location on the tool. In a number of cases, scanning electron microscopy (SEM) and electron probe microanalysis (EPMA) were used to examine the microstructure and microchemistry. A strong emphasis was placed on the microstructural consistency of the steels and irons throughout the collection. We looked for changes in fabrication and processing technology and in mechanical properties of the tools.

The saws examined in this study were martensitic and were very consistent in terms of microstructure and hardness, with the exception of the Manwaring saw (S2) which had superior hardness. Like the saw group, the plane blade group was found to be very consistent in terms microstructure and hardness. The plane blades examined were constructed by forge welding a steel edge onto a lower carbon wrought iron body. The tool was then annealed and quench hardened to produce a martensitic structure at the tip, and them tempered. Similarly, all the chisels examined in this study (mortising, tanged and socket) contained a steel edge forge welded to a lower carbon wrought iron body. In most cases the steel edge was a fully martensitic banded structure indicating the presence of phosphorus segregation, and the body was fully ferritic. The major exceptions were two tools (MC13, TC4) which had a more uniform martensitic steel edge. MC13 was stamped "cast steel" and TC4 was stamped "CS", which could be an indication that the steel edges of these chisels were constructed from a cast steel rather than a more inhomogenous but less expensive blister steel.

Introduction

Not unlike other cultures that have established colonies in far away lands, the European peoples who migrated to what is now the US brought their culture with them. Their way of life changed little as they became members of the New World. Colonial American cities were like replicas of the homeland. Echoes of Britain, Germany and Ireland could be heard if one listened closely. Socially, people interacted as they did in Europe. Even the new political system had similarities to those of Europe.

Just as the colonists brought their culture and ideals, they brought with them their trades and associated technology. Of the utmost importance to any colonial tradesman were his tools or implements which were at the heart of his business. This diffusion of technology from Europe to the American colonies was an important aspects of our work. Therefore, we began our study by examining the toolmaking industry of eighteenth century England.

There were four main centers for toolmaking in England at this time. Sheffield, Birmingham, London and the Lancashire area each specialized in the production of certain types of tools. In 1806, Great Britian alone produced 258,000 tons (TYL '80) of iron which would then sell for about £18 2s per ton. In 1840, the annual export value of tools and cutlery from England was estimated to be around £1.8 million (TYL '80) which accounted for less than half of the total steel produced. Ore consumption during this time was very localized due to problems with transport. As newer and more effective means of transportation developed, metalworkers were able to obtain a wider variety of ores. Some ores were high in phosphorus and were not suitable for use in the developing steel industries due to the embrittling effect that phosphorus has on steel. But of course this was not understood at the time. As the availability of ores increased, the local steel industries were able to select ores which would improve the quality of their steels and irons. At the time it was understood that certain ores were appropriate for certain applications. This new versatility in available ores allowed smithing operations to tailor-make a superior product.

When the American colonists needed tools for work, they normally bought imported British wares from local merchants. Much like with their other supplies, the colonists carried on trade with their motherland in order to obtain the tools necessary for their trades and personal projects. This created a transfer of objects and technological expertise from England to the newly established American colonies. As the colonists became more established they desired to manufacture their own tools, and when they lacked expertise, master iron makers and smiths were brought over from Europe, either as contracted employees or as new immigrants. Large iron works and mills were built in response to the new society's demand for implements. In 1650 the large Saugus iron works was established near Boston, MA. The Saugus enterprise was a large scale factory enterprise involving joint financing, complicated technology, specially imported workmen and heavy capital risk. In 1771, the Hopewell Furnace in Berks County, PA built its first cold blast charcoal furnace that produced pig iron and castings.

At the heart of the colonists technology were their tools. The implements by which craftsmen built their homes and conducted their trades appear to be simple to the modern eye. But upon closer examination it can be seen that these tools were developed only after many years of metalworking experience by highly skilled professional toolmakers. To highlight the importance of tool technology in the American colonies, the Colonial Williamsburg Museum recently mounted an exhibition entitled "By Their Smooth Handles" which specifically focused on woodworking tools (GAY '93).

Woodworking tools were the types of tools most commonly owned by the American colonists. As pointed out in the catalogue accompanying the woodworking tool exhibit, "They survive in far greater numbers than tools used in other manufacturing trades, and they are the types most frequently found at archaeological sites" (GAY '93). Thus by understanding these tools we might be able to understand more about the people who used them and their way of life. As mentioned above, these tools appear to be quite simple but actually they are "products of highly skilled, professional toolmakers". (GAY '93) Historical data tells us that these tools were keenly sold and distributed in America by English based makers. Apparently, "the marketing of imported tools was sophisticated and they were readily available to most Americans". (GAY '93) The use of these tools and other implements contributed greatly to the economic connection between England and America.

Another question commonly asked by Colonial American historians concerns the quality of these types of tools. "How could such wonderful buildings, chairs, chests-of-drawers, wagons and boats be made with the "crude implements" of two centuries ago?" (GAY '93) The technology of two centuries ago is not as primitive as one would suspect. Massive industrial complexes, automobiles and microwave ovens may not have been in existence, but the relative technological development of seventeenth and eighteenth century Europe when compared to the thousands of years of man's previous existence is rather sophisticated. From this holistic standpoint, two centuries ago was not that far back in history. Each tool in use at the time was designed for a specific purpose just as the tools of today. As different types of jobs evolved so the necessary tools required to make a job easier evolved. The quality of the tools coupled with the ingenuity of the craftsman resulted in large scale production and availability of many new household goods and contributed to the technological base for the Industrial Revolution in the nineteenth century.

In order to provide some insight into these questions concerning the identity and quality of woodworking tools, the Colonial Williamsburg museum sponsored this present study. A collection of tools from the exhibit at the Colonial Williamsburg Museum

(1994) was brought to Lehigh University for in depth metallurgical examination. The tools were subjected to detailed metallographic analysis which emphasized light optical microscopy (LOM) of both the macrostructure and microstructure and to microhardness measurements as a function of location on the tool. In a number of cases, scanning electron microscopy (SEM) and electron probe micro analysis (EPMA) were used to examine the microstructure and microchemistry. The museum selected a total of 33 eighteenth century woodworking tools for our consideration.

This group was made up of 6 saws, 6 plane blades, 14 mortising chisels, 3 socket chisels and 4 tanged chisels (Table 1). Each tool was compared not only to the other tools of its type but to all of the tools in the entire group. A strong emphasis was placed on the microstructural consistency of the steels and irons throughout the collection. Within the time periods these tools represented, we looked for changes in fabrication, processing and mechanical properties of the tools examined (Appendix A).

The saw group included saws made from several different types of steels that were in common use during the late eighteenth and early nineteenth centuries (Table 2) (Fig 1). Our testing objective for this group was to determine the general nature of saw blade materials, their fabrication and the hardening and tempering processes that they underwent. Also we compared the London (S2 and S5), Birmingham (S1) and Sheffield (S3, S4 and S6) examples for any significant differences.

The plane blade group (Table 3) (Fig 2) consists of five blades from London which probably date to 1740 - 1830 and one from Massachusetts which was probably American made. Of the five London blades, only one had a maker's mark which could serve as a positive identification. The museum believes that the earlier blades are either from London or Birmingham while the later blades could have originated in Sheffield. Our objective here was to determine the general nature of the plane blades with an emphasis on the materials, their fabrication and their hardening/tempering procedures over the 1740 - 1830 period.

Among the three types of chisels (Table 4) (Fig 3), the mortising chisels were by far the largest group. MC1 and MC2 are two mid eighteenth century Birmingham examples. MC3 - MC5 were all marketed by the firm of Thomas or Samuel Newbould circa 1770 - 1820. MC6 is a late eighteenth / early nineteenth century chisel that is qualitatively comparable to the Newbould group. MC7 - MC10 are four Green chisels that were sold circa 1800. MC11 is probably a much later example while MC12 and MC13 are nineteenth century chisels. MC13 is of particular interest because it has "cast steel" stamped on its body. MC14 is a late eighteenth / early nineteenth century chisel which has a broken tang. Our objective with this group was to determine the general nature of mortising chisel materials, their construction and subsequent processing with the following questions in mind:

1) What was the nature of the mid-eighteenth century examples manufactured in Birmingham (MC1 and MC2)?

2) Are there differences in these Birmingham (MC1 and MC2) chisels and early Sheffield (MC3 - MC5) examples?

3) Does the range of Newbould chisels reflect changes in Sheffield products during the late eighteenth / early nineteenth century (MC3 - MC5)?

4) How consistent are the chisels marketed by one manufacturer (MC7 - MC10)?

5) How do the roughly contemporary Law and Newbould (MC3 - MC6 and MC11) chisels compare to the Greens chisels (MC7 - MC10)?

6) How does the early nineteenth century Law chisel (MC6) compare to earlier examples (MC1 and MC2)?

7) What is the nature of the material in the un specified *cam* chisel (MC12)?

8) What is the nature of the cast steel in the so marked *cam* chisel (MC13)?

9) What can be discovered about the reason for the broken tang on the Weldon piece (MC14)?

Finally, the socket and tanged chisels were also studied in order to understand the materials, their fabrication and their processing (Table 5) (Figs 4, 5). We desired to better understand the nature of steel in mid century Birmingham chisels (TC1 and SC1) and compare the Sheffield examples (TC2 - TC4, SC2 and SC3) to the remaining Birmingham examples (TC1 and SC1). Because tanged chisels are relatively light duty tools when compared to the heavy duty socket chisels, it was our aim to compare the differences in handling of fabrication, materials and materials processing between these two types of chisels. The tool termed TC4 is stamped "CS" which could mean cast steel or simply refer to someones initials. We desired to find some evidence that might help us understand the meaning of this "CS" stamp.

Processing and Manufacturing

In order to provide the reader with some details concerning the technological problems encountered by toolmakers and the nature of the raw materials with which they worked, this section outlines the historical development of processing and specific terminology for different types of iron, steel and brass.

I) Iron and Steel

Steel is an alloy consisting of the metallic element iron and at least one addition - carbon, at a percentage of between 0.2 and 1.2 weight percent. The earliest furnaces developed by mankind were not capable of reaching a temperature high enough to make this material directly, but many cultures developed a method to make a crude low carbon *spongy iron* or *bloomery iron* which when hammered produced a more densified *wrought iron.* Wrought iron is tough in that it can absorb impact without breaking but is not very strong or hard. If thin pieces of wrought iron are held in a charcoal furnace for times varying from hours to days, carbon is absorbed into the iron and results in the foundations of steel. This solid state carburization process was known as cementation. If the steel is reheated and then rapidly cooled (quenched) in water or some other liquid medium it can be made much harder because the carbon is retained in a supersaturated crystal structure known as *martensite*. If the steel is then held at a mild heat for short periods of time (*tempering*) its properties can be further modified. As the martensite structure is modified in this manner the carbon separates out as distributed carbides. After producing bloomery iron, it was usually desirable to carburize the metal which resulted in steel. In eighteenth century England this was normally done by one of three methods.

One method for carburization involved stacked layers of bar iron which were placed in a sealed chest with powdered charcoal for very long times. The resulting product was termed *blister steel.* Blister steel was produced from iron bars that were about 2 - 3 in wide and 1/2 - 3/4 in thick. Once inside the sealed chest together with charcoal the temperature of the chests were brought up to about 1050 - 1100°C and held there for 5 - 7 days. This blister steel which was produced by the cementation process derives its name from the blistered appearance of the surface of the steel after its long heat treatment.

The second process which was developed became the typical method for making steel in Sheffield. It involved remelting blister steel in large crucibles and then casting it into molds. These castings could weight anywhere from 13 - 70 lbs. This process provided a more uniform material unlike the result of the previously used blister steel. Because blister steel was carburized from the surface inward, which caused an inhomogenity in carbon content to develop. The outer areas of the bars were much higher in carbon content than the interior of the bar. The process for making *crucible steel* began by breaking up bars of blister steel and placing them into large crucibles which were then sealed. The crucible was then covered with coke and heated to around 1600°C. This temperature was much higher than had been attained and in use prior to the advent of crucible steel. The crucibles were preheated and specially constructed to withstand these new thermal conditions. After heating for several hours until the blister steel had become liquid, the slag was skimmed off and the metal poured into ingot molds. This crucible steel was therefore more homogenous and had fewer slag inclusions than the blister steel from which it was made. Due to the higher quality of crucible steel it was considerably more expensive.

Another type of steel which was commonly made in Sheffield was called *shear steel*. Shear steel was used in cutlery and for edged tools such as the ones examined in this work. To make shear steel the previously made blister steel had to be examined by a skilled smith to determine its relative carbon content. The discovery and identification of carbon as an element and therefore the modern understanding of the difference between bloomery iron, cast iron and steel, did not occur until 1821 (SMI '81) (BRE '63). The smith could only use the fracture surface appearance in order to provide a loose correlation between appearance and quality. Also in these processes other alloying elements such as Ni, Cr, V, Mo, etc. which improve properties can also be added. These bars of blister steel were then rolled into thinner bars and stacked on top of each other according to their carbon content. The stacks or bundles of steel were then heated and forged together to form a sort of composite with layers of slightly differing carbon content. The layers of high and low carbon steel resulted in a product which had superior qualities. Blades made from this material could hold a good edge and were flexible.

As people learned to build furnaces with higher temperature capability by using a draft or blast furnace, it became possible to melt and cast iron directly. However this *cast iron* contained too much carbon (usually about 4%) and had to be processed further due to its brittleness. Cast iron is sometimes called *pig iron* because the liquid iron was allowed to run off into the casting floor in a shape of a larger central ingot with many smaller ingots running perpendicular to it, reminiscent of a mother pig suckling its young. The blast furnace was believed to be first introduced in western Europe around the fifteenth century. (TYL '80) It had advantages over the previously used bloomery process in that almost one ton of liquid ferrous metal of high carbon content could be produced at one time where previously only small "blooms" of low carbon content iron could be produced from the bloomery process. This new process was believed to be first introduced in England by the French to the Weld of Sussex and Kent around the end of the seventeenth century. (TYLE '80)

As the technology for the preparation of high temperature refractories and for achieving higher temperatures in furnaces of significant size improved, new processes developed for converting the higher carbon cast iron back to steel. In 1856 Henry Bessemer in England and Kelly in the US invented the Bessemer Conversion process (BAR '81a). This process uses a tiltable refractory lined iron vessel into which liquid high carbon iron from the blast furnace is poured. The Bessemer vessel is then tilted to a different position and air is blown through the liquid iron preferentially oxidizing the carbon out of the iron to produce *Bessemer Steel.* When this process was first used, there was a patent fight between Bessemer and Kelly which Bessemer won even though Kelly had conceptualized the process earlier.

Open hearth steel was produced by pouring the liquid high carbon iron from a blast furnace into a furnace with a large flat hearth. Hot air was passed over the liquid iron bed to preferentially oxidize the carbon out of the liquid high carbon iron to produce steel. This became the predominant method worldwide for steel production and was in use until just a few decades ago. Currently, most steel is produced by the Basic Oxygen Furnace which uses a vessel similar to the Bessemer Converter but has a lance that is pushed into the iron through which pure oxygen (rather than air) is blown. The process is exothermic, economical and produces high quality *Basic Oxygen Steel.*

All steels produced via a liquid process step (crucible steel, Bessemer steel, etc.) have much better separation of slag and therefore yield more uniform, higher quality steels. Also, phosphorus content is more easily controlled by these processes and they usually guaranteed steels of good hot ductility and hardenibility.

II) Manufacturing of Edged Tools

Today, because of the relatively low price of steel, an edged tool is often made from a single piece of steel with only the cutting edge quench hardened and tempered. Until recent times, because of the high cost of quality steel, the body of the tool often consisted of a less expensive (but softer and tougher) low carbon iron body onto which was attached a more expensive edge which was made of a harder, higher quality steel. (Fig 6) This configuration allowed the bulk of the tool to be constructed from a less expensive material such as iron or a very low carbon steel while optimizing only the cutting edge. Most of the chisels (mortising, tanged and socket) and plane blades were constructed in this manner. The steel edges vary in size and shape depending on the function of the tool. The multistep process explaining the general steps for fabrication of an edged tool in the 18th and 19th centuries is as follows: (HAR '94) (RIC '78)

The iron which will become the body of the tool and the steel used for the cutting edge were kept in the same smithing furnace and fluxed with sand or borax until the welding temperature was reached. Next, the body of the tool was formed and the steel was forge welded on to the cutting edge. The composite piece is then buried in wood ashes at about 1500°F (a cherry red heat) (UNT '80) and allowed to slowly cool for about one day. This acted as an annealing step which relaxes strains introduced during forging. We also believe that this step was necessary to complete a good bond between the iron body and the steel edge. At this temperature, carbon was able to diffuse from the steel into the iron so a better metallurgical interface at the lamination would result. Next, the oxide scale or FeO that developed on the surface of the tool during the slow anneal was filed or ground off. Then the tool was again heated to a cherry red heat, approximately 1500°F and water quenched. This process was designed to harden the steel. Next, to alleviate the problem of brittleness in the quench hardened steel, the piece had to be tempered. This was done by heating a separate block of iron to a yellow heat color which corresponds to a temperature range of 1260 - 1371°C (ANO '67). The chisel was placed on the block with the steel side facing up. The temper operation continued until a thin surface oxide having a purple color (260 - 304°C) (UNT '80) popped out on the steel edge. Finally the piece was water quenched to freeze the temper.(Fig 7)

III) Brass

According to the Williamsburg museum (GAY '90), saws were among the most technically complex woodworking tools to manufacture. Compared to other types of woodworking tools, saws have the most complicated set of mechanical design considerations. They need to be hard, smooth and flexible in order to perform properly. The group we examined included four dovetail saws and two sash saws. All of the saws had a reinforced backing fashioned out of brass with the exception of S4 which had a backing made of iron. This type of opposite edge backing provided more support and stiffness for the blade which made the saw more stable during use.

Typically a brass backing was installed by mechanically clamping a brass sheet around the top of the steel blade. The brass used for these backings was probably either stamping brass (67 / 33) or yellow brass (70 / 30) (TYL '80). The specific details of the brass piece that were examined will be left for later sections, but a brief discussion of the history and manufacturing techniques of brass is in order.

First copper must be extracted from its ore. Several different methods for copper production were used in England during this time. One such method for copper smelting in Bristol is as follows:

(TYL '80) First the furnaces were preheated for one to two days and the ores were calcined if necessary. 300 lbs of ore (usually mixed copper oxides, and sulfides with high iron content) were loaded into the furnace at one time with a flux of lime with silica. Sometimes slag from previous smelting operations was also added. "The furnaces were recharged every four hours and the slag was tapped every twelve hours." The resulting product was poured into sand molds. This enriched sulfuric copper product was known as *matte* and was then broken up and roasted with coke and coal for 12 - 14 hours in order to convert the copper sulfide to copper oxide. The temperature was then increased to melt the copper oxide which was poured into sand molds. This matte was again broken up and roasted with coke and coal. This roasting and melting process was repeated indefinitely until the desired grade of copper was produced. After this, more refining was carried on in separate furnaces.

To make brass, one needs to add to copper the alloying element zinc. Most zinc used in Europe during the 18th and 19th centuries was imported from either India or China. As knowledge of the zinc smelting technology became better understood, it was then produced in Europe as well. Zinc can be obtained from only a few different ore sources. (TYL '80) Sphalerite, or zinc sulfide (ZnS) was probably the most abundant ore available locally in England during the 18th and 19th centuries. But Smithsonite, which is a carbonate of zinc, is much more easily smelted. Usually broken pieces of the above described copper are layered with crushed and calcined Smithsonite ($ZnCO_3$) in large crucibles. Ideally the temperature should stay below the melting temperature of copper which is 1084.5°C. But the zinc needs to be reduced and vaporized which occurs above 903°C. (Fig 8) (ANO '72) This temperature range must be maintained for a long enough time so that the zinc vapor can be absorbed by the copper pieces. The resulting product is then rolled, drawn or processed as necessary. It should also be noted that in the 18th and 19th centuries, different compositions of brass were developed for different purposes, just as alloys are designed today.

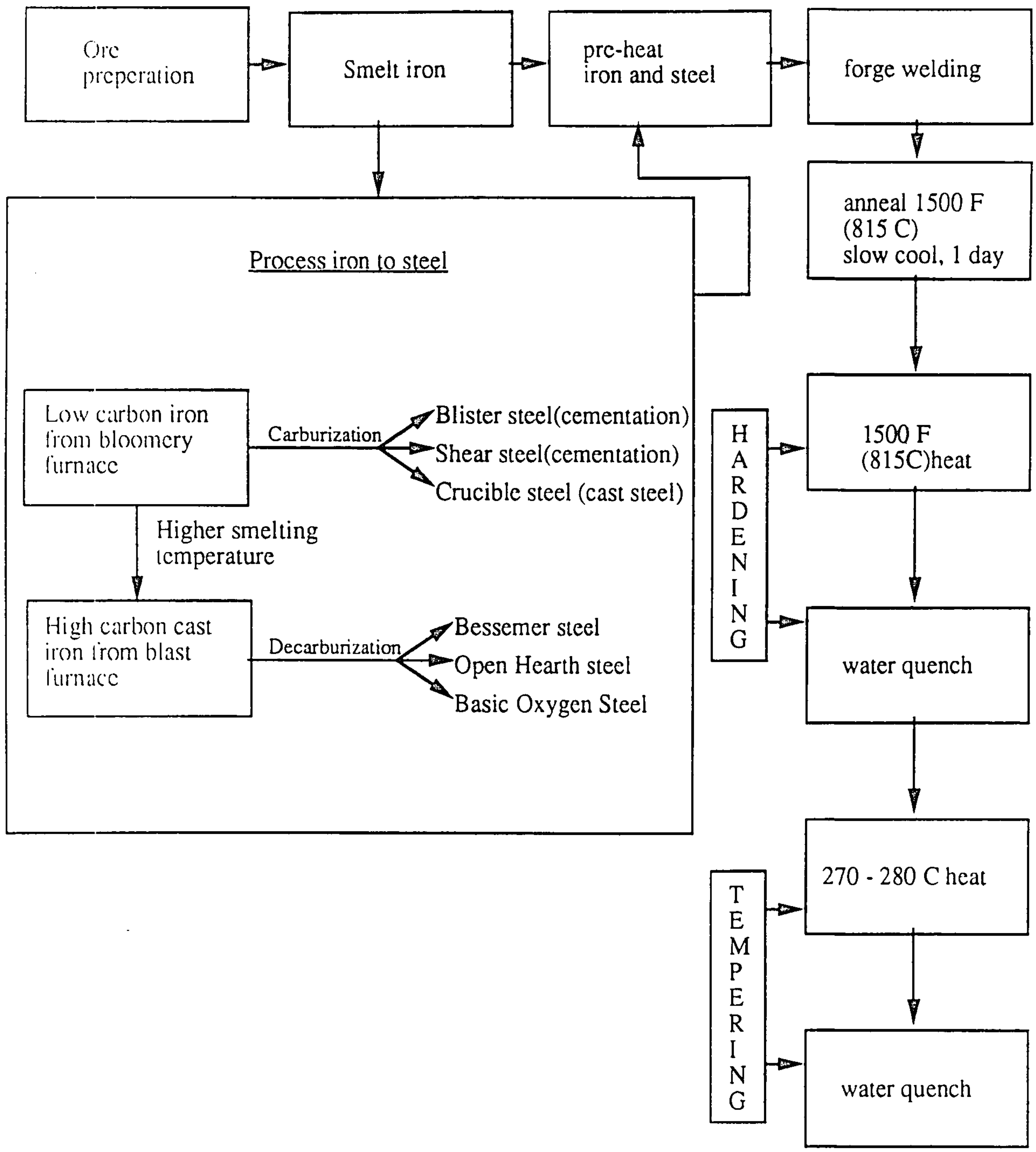

Ore preperation
Smelt iron
pre-heat iron and steel
forge welding
anneal 1500 F (815 C) slow cool, 1 day
Process iron to steel
Low carbon iron from bloomery furnace
Carburization
Blister steel(cementation)
Shear steel(cementation)
Crucible steel (cast steel)
Higher smelting temperature
High carbon cast iron from blast furnace
Decarburization
Bessemer steel
Open Hearth steel
Basic Oxygen Steel
HARDENING
1500 F (815C)heat
water quench
TEMPERING
270 - 280 C heat
water quench

Figure 1

Experimental Procedure

I) Specimen Preparation

Metallographic examination of a metal sample usually consists of the destructive analysis of a portion of the object which is mounted in some medium to hold it while it is polished with a succeedingly finer array of polishing powders. The specimen is then etched to develop the fine details of the microstructure. This destructive method usually gives the best results in terms of resolution and clarity of the image produced. However, archaeometallurgists often find themselves in situations where they are not permitted to destroy the sample and are therefore required to perform nondestructive in-situ testing on metallurgical specimens. In-situ examination can be inconvenient and trying while at other times such as with this Williamsburg handtool collection, nondestructive testing can often be much simpler. Sometimes, if direct in-situ examination does not give a good result, it is possible to prepare a replica of the surface and then examine the replica rather than the real specimen. All three methods (in-situ, replication and destructive testing) were used in the present study depending on the specific specimen, but in-situ examination was the most commonly utilized method.

II) In-situ Preparation

Our procedure developed for the requested in-situ metallography involved a Sturers movipool - 130 portable polishing unit. This unit was able to metallographically prepare a surface in situ. Every single specimen be it steel, wrought iron or brass was polished, etched and optically examined in-situ. The mortising chisels were in situ polished on the entire bottom side of the blade which revealed an area of wrought iron, an area of steel and an interfacial area between the two. Similar interfacial structures were also seen later in the plane blades, socket chisels and tanged chisels. All iron and steel specimens that were in-situ polished with the Sturers movipool - 130 were taken through the following grinding and polishing sequence:

400 grit SiC

600 grit SiC

6μm diamond

3μm diamond

.3μm Al_2O_3

.05μm SiO_2

This provided a scratch free surface that was ready to be etched and examined. The preparation of the brass backing on the saws was only slightly different in that prior to the final etching step included an attack polish (a light etching solution was used during the polishing step) of ferric nitride and .05μm SiO_2 powder. This removed scratches left by the .3μm Al_2O_3 and began to slightly etch the surface.

After polishing was complete, an etching procedure had to be developed. There were several types of etches used for this project. The type of etch used is noted by each photograph. As just mentioned, the brasses received an attack polish with .05μm silica powder with ferric nitride and were then etched with ammonium hydroxide, water and 3% H_2O_2 that was swabbed on with cotton and a glass rod. The iron and steel specimens first received a 4% picral etch and then were usually followed by harsher etches such as 2% nital and/or Marshall's etch (5 ml H_2SO_4, 8 g oxalic acid and 100 ml water) depending on the specific situation. Also some of the steel and iron specimens which were thought to contain phosphorus were etched with Stead's reagent (10 ml H_2O, 900 ml alcohol, 10 g $CuCl_2$, 20 ml HCl and 40 g $MgCl_2$) which deposits copper on areas of low phosphorus or with Oberhoffer's etch (500 ml H_2O, 500 ml C_2H_5OH, 42 ml HCl, 30 g $FeCl_2$ and .5 g

$SnCl_2$) which also delineates phosphorus segregation in steels. Again, all in-situ etching was done by carefully swabbing the etch onto the exposed metal area. This technique provided more control than submerging the entire specimen in the etchant.

III) Light Optical Microscopy (LOM)

Immediately after metallographic preparation, all specimens were placed into a desiccator for environmental protection against oxidation of the polished surface. The next step was to observe the specimens in-situ under a light optical scope. The procedures for mounting the tools under the microscope required that the surface being observed be perfectly flat. Normally the tool in question was placed on top of some plasticine which was then mounted on a glass slide. At times, two or more glass slides with plasticine were used for balance. The tools were then adjusted to maximum possible flatness by positioning them where the most amount of light would be reflected back into the microscope.

IV) Replication

Unfortunately, it is very complicated to get large specimens which were in situ polished perfectly flat under a microscope. Therefore, many of our in-situ photographs were somewhat out of focus. To remedy this situation we decided to take replicas of the relevant areas for further study in order to obtain better in-focus photographs. This also allowed us to retain a permanent record of the microstructures that would be small enough to view easily and was convenient for storage.

The replication process required a polished and etched surface. After polishing and etching the surface was wetted with acetone and a piece of .88 mil replicating tape (cellulose acetate) was placed over the area. The tape was partially dissolved by the acetone so that it conformed perfectly to the morphology of the surface. After about 10 minutes the tape was dry and could be pulled off with tweezers. Then the tape was laid on a glass slide and was sputter coated with Au-Pd in order to obtain better contrast in the microscope.

This replication process provided a reliable means for examining microstructures conveniently.

V) Destructive Sampling Procedures and Sacrificial Specimens

Several pieces from each tool group were selected for destructive examination so we were able to cut very small specimens from the parent piece. Because many of the tools could not have a section cut from them, we tried to only cut pieces from those tools which we thought would represent others of its type. These cut pieces were then mounted, metallographically prepared, photographed and in some cases were also examined in the scanning electron microscope (SEM) for higher magnification images than what can be obtained from LOM and for Xray microanalysis.

The electron beam generates two types of images. An image produced from the beam current, called a secondary electron image (SE image) and one by electrons scattered back into the direction of the electron beam (BSE image). The BSE image is sensitive to the average atomic number of the point under the beam. The electron beam scans the surface being studied and also generates Xrays from each point under the beam. These Xrays are of a specific wavelength characteristic of the elements present locally under the beam. Thus three separate output signals are used in SEM study.

Pieces were cut from the following tools in the described manner and specific locations:

S2 --> Removal of the back tooth near the handle.

S3 --> Removal of about 1/8" X 3/8" of brass backing which included a similar sized piece of the steel blade or sheet from underneath the handle area.

S6 --> Removal of the back tooth near the handle.

MC1, MC2, MC7, MC12, MC13 --> Removal of a 2mm slice parallel to the current sharpening bevel.

TC4 --> Removal of approx. 2mm X 1mm of the steel tip.

SC1 --> Removal of a 2mm slice from the edge and another similar piece from the socket end.

** Note that TC2 and SC3 were dropped from our examination due to heavy corrosion and scale.

In addition to these cut specimens, the Williamsburg museum made two sacrificial pieces available to us : one plane blade (MPS) and one mortising chisel (MC14). This allowed us to fully examine and understand these types of tools because we were able to cut and examine them as necessary. The steel edge of the sacrificial plane blade was metallographically examined thoroughly. A piece cut from the steel edge of MPS was austenitized at 950°C for 20 minutes and allowed to slowly cool in an attempt to develop an identifiable structure under known conditions. The sacrificial mortising chisel was sliced in half lengthwise and subjected to a macroetch (HF, HNO_3 and water). This etch revealed features such as small cracks, flow lines and the steel cutting edge. This enabled us to better understand the forming processes of the chisels and other tools which were made in a similar fashion. Complete descriptions of both of these pieces are included in the text within their respective groups.

VI) Microhardness

Once an understanding of the microstructures was obtained, the next step was to obtain some idea of how these tools behaved mechanically. A good quantitative but macroscopically nondestructive test for understanding mechanical behavior is microhardness. All the tools included in this report were subjected to in-situ microhardness tests. This allowed us to verify and interpret our metallographic data and gave us a good basis for comparison. We were able to examine the tools not only within specific groups but also from a more holistic standpoint where we compared tools of different types. The indents we made were not visible with the naked eye so we had no need to worry about defacing the tools. The only problem was how to mount the large tools under the indentor.

Various techniques were used depending on the size and shape of the tool in question. The smaller tools were mounted on plasticine and placed on glass slides and positioned under the indentor. The larger tools required a combination of plasticine, glass slides and even C-clamps for the proper positioning.

A Vickers Microhardness indentor with a 200 gram load and a 15 second dwell time were used on all specimens. The in-situ microhardness tests that we could make were limited in that we were only able to take a certain amount of measurements, prescribed by the Colonial Williamsburg museum. But the cut and sacrificial pieces were available for unlimited measurements. These specimens often provided a stability check against the in-situ measurements which could have been erroneous if not mounted properly in the plasticine.

Results

We all know how frustrating work can be if we attempt to use the wrong tool. Such was probably the case of MC14, the broken tang Weldon piece. These types of fractures were not uncommon in early America (GAY '92). We can hypothesize that the piece was probably improperly used as a prying tool but this is really beyond the scope of our investigation. Although, future researchers might wish to examine a group similar to ours with these kinds of questions in mind. There are several methods for non destructive analysis of materials such as liquid penetrant inspection, ultrasonic inspection or eddy current. Details of these and other nondestructive testing techniques can be found in the Metals Handbook, 9th ed., Vol.17 entitled "Nondestructive Evaluation and Quality Control".

I) Mortising Chisels

A) LOM

This group consists of fourteen chisels (Table 4), one of which (MC14) was literally cut in half lengthwise and subjected to a macroetch. This chisel, MC14, had a wrought iron body and a steel sheet which was forge welded in place along the cutting edge of the tool. This steel edge provided a superior cutting surface at a minimal cost. Instead of making the entire tool out of expensive steel which would provide a stronger, harder chisel it was only necessary to steel the cutting edge. The macroetchant revealed flow lines within the iron body and the precise location and size of the steel edge.(Fig 6) Also two cracks were found running perpendicular to the length of the blade.(Fig 6) On first inspection these cracks are indicative of a repaired fracture. In other words this tool fractured and was later repaired by forge welding the partially separated sections back together. Based on the location and microstructure of MC14, we believe that it was probably used as some type of prying tool when it broke.(GAY '92) The crack runs transverse through a region of higher carbon iron and is surrounded by slip lines.(Fig 6, 9) This method of repair was probably successful in that there were no signs of subsequent fracture.

Photomicrographs were also taken at various positions along the blade of MC14, some in the steel edge and others in the wrought iron body. Although termed wrought iron, the material used to form the body is really an imhomogenous low carbon steel. Neumann bands indicating a cold worked structure were also observed in the wrought iron body along the outer top edge of the blade.(Fig 6, 10) The steel edge had fine lath martensitic structure and a homogenous single phase glassy iron silicate slag.(Fig 11) The slag inclusions in the wrought iron body were embedded within ferritic grains and were elongated in the direction of forging.(Fig 12) They consisted of a large dendritic second phase which was identified as wustite or FeO, a crystalline iron silicate, the grayish matrix surrounding the dendritic phase was mostly fayalite ($FeSiO_3$). The very dark (almost black) areas in the matrix of the slag are a lower atomic number residual glass phase that was left uncrystallized. The wrought iron also contains smaller, spherical slag inclusions similar to that observed in the steel. The spherical inclusions in the wrought iron area of MC14 resemble in color, morphology and shape the single phase slag inclusions identified in other specimens such as MC12 and SC1-socket which were confirmed by microprobe analysis to indeed be a iron silicate.(Fig 13, 14, 36)

Finally, the interfacial contact during forging and annealing between the high carbon steel edge and the wrought iron body of MC14 allowed the solid state diffusion of carbon from the higher carbon steel edge into the lower carbon wrought iron body.(Fig 16) The carbon diffusion created a gradation in microstructure depending on the relative amount of carbon present. Just on the wrought iron side of the interface, an area of larger lath martensite developed where the prior austenite grain boundaries are sometimes visible.(Fig 17) Gradually as carbon content decreased while moving further into the iron section, large ferritic grains begin to develop.(Fig 16) This gradation in microstructure or some variation of it has been observed in all of the other edged tools near the forge welded area. The ferrite grains were not a uniform size throughout the piece due to local differences in carbon content, uneven temperatures during forging and other impurities.

A general metallographic description of the mortising chisels is as follows. The remainder of the chisels were polished in situ along the bottom edge of the body. Subsequent etching revealed the steel edge, the wrought iron body and the interface between the two. The steel edge always appears as a tempered martensitic structure with some retained austenite.(Fig 18) The steel edge shows a banded structure similar to that seen in the other tool groups. Upon etching with Stead's etch this banding could be identified as being related to segregation of phosphorous during smelting.(Fig 19) Stead's etch deposits copper on high phosphorus regions of the steel. The dark bands were low in phosphorous and the light bands were high in phosphorous. The slag inclusions in the steel is usually elongated, single phase and quite small when compared to slag in the wrought iron body. Typically the wrought iron body of the chisels in this group contained ferrite grains and elongated two phase slag stringers as described earlier. Most of the chisels also contained spherical single phase slag inclusions inside the ferrite grains.(Fig 13) The structures found in the wrought iron region of the chisels varied from chisel to chisel depending mainly on the local carbon content. Some ferrite areas also contained *pearlite* while other areas contained a meandering line within the ferrite, sometimes known as *epitaxial ferrite or halo ferrite.*(Fig 20) This epitaxial ferrite was attributed to growth of new ferrite on previously existing undissolved ferrite as the specimen cools from the ferrite + austenite region. (SHR '66)

B) Hardness

Microhardness tests were performed in situ on all the mortising chisels and on the cut specimens.(Fig 49) These results were used to obtain a quantitative basis which would provide a good method for comparison for all the tool groups. In this manner, a hardness profile was constructed which reflected changes in microstructure as the indentor moved from the steel edge into the softer wrought iron body. Thus we were able to confirm our previous interpretation of the metallographic structures and compare the relative properties of the tools.

As can be seen from Figures 21-23, most of the profiles had a very low hardness of about 100 - 200 Hv in the wrought iron bodies. As the indentor approached the interface a sharp increase in hardness was typically observed. This was the transitional region between the steel edge and the wrought iron body. Hardness dropped from its maximum of around 800 Hv to its minimum of about 150 Hv in a space of about 2.25 mm. Finally as the indentor moved into the steel edge hardness values began to level off again at between 700 - 800 Hv. This corresponds well to the accepted hardness of martensite which is normally 390 - 800 Hv depending on carbon content and tempering condition.(ANO '78)

For MC1, MC2, MC7, MC12 and MC13 additional hardness tests were performed on the cut and mounted samples to compare with and confirm the in situ results. As can be seen in Figures 21, 22, 27, 32 and 33, these data points corresponded quite well with the previously performed in-situ experiments.

C) Microchemical analysis

Due to the common features observed in the single phase slag inclusions in the steel edge and the multiphase structures seen in the slag inclusions in the wrought iron body of the chisels, plane blades, Xray microchemical analysis in the SEM was limited to only a few representative specimens. There were two types of slag inclusions that were observed in the mortising chisels. The first was a single phase slag which was seen only in the steel edges of the chisels. These slag inclusions contained Si, Fe, Ca, K, Mg, Al and Mn, Ti.(Fig 34) The two phase slag seen in the wrought iron body of the other mortising chisels consisted of a matrix that also contained Si, Ca, Fe, K, P, Mg, Al and Mn.(Fig 35)

The lighter dendritic second phase is wustitie or FeO. (UNG '87,'90,'91) The matrix of this slag contained both dark and light areas. The dark areas were a crystallized glassy portion of the surrounding lighter gray uncrystallized fayalitic matrix. (BAC '82).

The cut specimen from MC12 showed a larger than normal amount of spherical slag inclusions in the body. These inclusions were observed only in the wrought iron section of most of the specimens examined not only in this group but in the other chisel groups as well. The identity of these inclusions were confirmed by an Xray microprobe EDS scan on the similar spherical inclusions present in SC1 which was then compared to the previously taken spectra of the larger two phase elongated slags. These two spectra show the same relative amounts of the same elements. (Fig 14, 35, 36)

D) Anomalies

An exception to the typical interfacial area observed in most of the mortising chisels was MC9 which seemed to have a rather large "intermediate" zone of approximately 10.2 mm which could be seen with the naked eye. This tool contained three distinct regions along the bottom surface of the tool, two of which were in the "wrought iron" section of the body. The first area was the steel edge. The second and third areas were the "intermediate" zone and the remainder of the body. These second and third areas consisted of pearlite and cementite. The "intermediate" zone contained more cementite and small amounts of pearlite (Fig 37) while the final zone was almost all pearlite with cementite lining the grain boundaries (Fig 38). This was reflected in the hardness profile of MC9. Notice that it had a much more gradual drop in hardness with distance than the hardness profiles for the other mortising chisels.(Fig 29) The only tools without an obvious steel edges were MC4, MC5, MC8, MC9 and MC10. The hardness profiles for these tools indicated some type of higher hardness material located at the steel edge. But our metallographic results did not indicate an interfacial region as seen in the other mortising chisels. This could possibly be due to the fact that the weld was clean but further study is needed to determine the manufacturing procedure for these tools.

MC2 had a banded structure in its steel edge with large veins of higher carbon running from the edge into the wrought iron body.(Fig 19, 39) The innermost region of these veins contained large laths of martensite which on either side have a very small "diffusion zone".(Fig 40) Because of their placement within the wrought iron body, carbon diffused from the vein into the surrounding ferrite and formed a small gradation in microstructure similar to that seen at the steel edge / wrought iron body interface. Also in the wrought iron region of MC2 were small spherical slag inclusions, elongated two phase slag inclusions and epitaxial ferrite or halo ferrite.

The wrought iron body of MC4 (a Newbold chisel) contained typical elongated two phase slag inclusions embedded in a ferrite matrix. The microstructures from the steel edge show pearlite and ferrite (Fig 41) while the body is ferritic.(Fig 42) This type of structure would be expected to have hardnesses much lower than the martensitic steels. This is confirmed by the hardness measurements taken in-situ from this chisel.(see hardness profile) The maximum hardness in the steel edge only reached about 400 Hv while the minimum hardness in the wrought iron body dropped to almost 100 Hv.

MC5 contained three distinct sections (Fig 43) across the top edge of the blade. The first section at the steel edge is a typical martensitic steel edge with high hardness of up to 700Hv.(Fig 44) The middle section contained pearlite and ferrite with hardness around 300 Hv. The third section nearest the handle contained mostly pearlite and single phase slag stringers, similar to the slag inclusions seen in the steels of other mortising chisels.(Fig 45) The hardness of the third section was not much different than the hardness of the middle section, i.e.. around 300 Hv.

The steel edge in MC7 had more of a feathery martensitic structure than some of the other steels in this group which is reflected by its high hardness of almost 800 Hv.(Fig 46)

The two phase slag inclusions in the wrought iron body didn't have very much of the dendritic phase, FeO. This indicated that the furnace conditions were more reducing than those for the bodies of the specimens which contained more dendritic FeO in the slag inclusions. Also the interfacial area of MC7 did have the large martensitic laths that were seen in the other specimens.

MC10, MC9, MC8 and MC5 are very similar in that the body of the tools contained both pearlite and ferrite.(Table 4) This area contained almost no slag inclusions but the slag that was present was single phase and there were no small spherical slag inclusions. They both had low end hardnesses of around 200 Hv. In contrast, their steel edges were very different. MC10's steel edge (Fig 47) had more of a feathery structure than the steel edge of MC9 while MC5 and MC8 had martensite in their steel edges and MC4 had pearlite and ferrite in the steel edge.(Fig 41) This was reflected in the hardness profiles in that MC10 only reached a maximum hardness of about 600 Hv while MC8 and MC9 soared up to around 800 Hv. MC5 reached about 640 Hv in its steel edge and MC4 was up to around 425 Hv in the steel edge.

MC12 had an abundance of small spherical slag inclusions in its wrought iron body.(Fig 13) It also contained larger, globular one phase slag inclusions. Its hardness profile appeared normal with a rather sharp increase in hardness with distance. MC13's steel region had a slightly lower hardness (700 Hv) than MC12 which was around 800 Hv. The steel edge of MC13 contained lath martensite with islands of ferrite, note that the prior austenite grain boundaries can be seen.(Fig 48) It's interfacial area did not have large lath martensite while the interfacial area of MC12 did. The wrought iron body of MC13 had ferrite grains with a combination of single phase and two phase slag inclusions.

II) Plane Blades

A) LOM

The plane blade group consisted of six plane blades (PB1 - PB6) and one additional plane blade designated sacrificial (MPS).(Table 3)(Fig 2) These pieces all contained a steel edge located on the bottom side of the plane blade which acted as a cutting edge. The remainder of the piece consisted of a wrought iron body. In-situ metallography revealed that in each and every case the wrought iron body of the planes consisted almost entirely of ferrite grains with elongated slag inclusions.(Fig 50) These slag inclusions were identical in appearance to those previously reported in the wrought iron sections of the mortising chisels. The steel edge had a tempered martensitic structure with two phase slag inclusions.(Fig 51) The interfacial region between the steel edge and the wrought iron body was similar in four of the plane blades (PB1, PB2 and PB4) and the sacrificial piece (MPS) while PB3, PB5 and PB6 contained a thin oxide layer at the interface which interfered with the formation of a true metallurgical bond.(Fig 53) In PB1, PB2 and PB4 carbon diffused from the steel edge into the wrought iron body which created a zone of varying carbon content. The steels edges in all of the planes were typically banded and contained small single phase elongated slag stringers. A quenched and tempered structure of lath martensite was common to all of the steel edges. In most cases there were oxide corrosion pockets which appeared in the wrought iron bodies of the tools. This was due to the fact that these results are from in situ micrographs and the structures seen are those on the outside region of the tool which had direct contact with the environment. This was not the case for the sacrificial piece because we were able to section it and observe the interior which was free from the deleterious effects of oxidation.

B) Hardness

There were three main areas of interest common to all the plane blades from this group. The first was the iron area which consists of the handle and the majority of the plane. This region need only be sturdy, reliable and easy to make. The second was the steel edge which was used for cutting and required a harder, more durable metal. The third was the interface region between the iron and steel areas. The interface region between the two metals (iron and steel) contained many clues relating to the nature of the metals and their interaction with each other. Therefore we made special effort to obtain hardness measurements from both the steel and iron areas as well as in the interfacial region.(Table 3)

Hardness tests were first performed on the sacrificial molding plane using a Vickers indentor with a load of 200 grams. The iron region, which consisted of mostly ferrite grains had an average hardness of 150 Hv and the steel region, which consisted of a tempered martensite structure has an average hardness of 579 Hv. This also agrees well with the standard hardness of martensite which is in the range 520 - 700 Hv. (ANO '78)

Tests performed on PB1 - PB6 yielded similar results. The iron bodies for this group showed hardness numbers ranging from 149 - 219 Hv while the steel bodies had hardness values of 586 - 609.5 Hv. We believe that the higher hardness values reported in the wrought iron bodies of PB1 (198 Hv) and PB5 (219 Hv) to be due to the presence of phosphorus in the wrought iron regions. (TYL '86)

More detailed hardness profiles were made on two of the planes : PB2 as representative of a tool showing a broad metallurgical interface and PB6 as representative of a tool with oxide at the joint interface. The hardness profile for PB2 showed a gradual increase in hardness from 159 Hv to 600 Hv over a distance of about 2.25 mm.(Fig 54) The hardness profile for PB6 showed a sharper increase in hardness from about 150 Hv to 580 Hv over a distance of only a few tenths of a mm, i.e.. the width of the oxide.(Fig 55)

C) Anomalies

1) The difference in hardness profiles between PB2 and PB6 has been described above.

2) In PB1 the hardness reading at the cutting edge was lower than in the interfacial region whose hardness was in turn unusually high; equal to the high hardness near the cutting edge of all other blades. The interfacial region of PB1 was in fact not straight but rather

had an obvious curvature along. Thus it was possible that the indentor moved into the steel region for the measurement recorded as the interfacial region and then back into the wrought iron area for the measurement recorded as the steel edge.

3) As mentioned previously, the microstructure within the steel edge of the sacrificial piece was unusual in that it consisted of martensite with islands of ferrite.(Fig 56) This structure would most commonly arise if a low carbon steel was heated to and quenched from the two phase region (ferrite and austenite) instead of being fully austenitized.(Fig 57) The significance of the observation was that the carbon content of the steel would be lower than expected.

In order to verify this point, a piece cut from the steel edge of the sacrificial piece (MPS3) was austenitized at 950°C for 20 minutes and then slowly cooled to room temperature while in the furnace. The resulting structure (Fig 58) consisted of pearlite and ferrite which verified our initial suspicion. The area fraction of pearlite was measured to be 48% on a light optical microscope equipped to perform quantitative metallography and using the Macintosh Ultimage 2.01 software (Fig 59). The small single phase slag inclusions present in the steel edge of MPS3 were confirmed to be an aluminosilicate by Xray EDS measurement.(Fig 60)

III) Saws

A) Blades (steel sheet)

1) LOM

This group was extremely interesting because the saws are actually composites made by clamping a brass backing onto a thin steel sheet or blade. The only exception to this was S4 which had a wrought iron back. In almost all cases (S1, S3 - S6) the sheet steel consisted of a tempered martensitic structure with finely dispersed small carbides which acted to toughen the steel. In S2, a coarser tempered carbide structure was noted. (Fig 61) The observed structures were very consistent throughout the group. There were no gradations in microstructure as seen in the other groups because the blades contain only one homogenous material, steel. There were hardly any observable slag inclusions in any of the saw blades. The small slag inclusions that were present in the steel blades were found to contain Si, Fe, Ca, Al, Mg, K, S, Mn and Ti by EDS.(Fig 63)

The backing of S4 was examined metallographically and was found to be mostly ferrite with small amounts of pearlite (Fig 62) and appeared to be made of a fairly low carbon wrought iron. Some small globular slag inclusions and large pools of corrosion scale were also observed.

2) Hardness

In all cases, in situ microhardness measurements were taken from three different locations along the length of the blade and also in an area of the backing material.(Table 2) In addition, two microhardness measurements were made on small specimens which were cut from the blade material of S2, S3 and S6. In all cases there was good agreement between the hardness measurements taken in situ and the hardness measurements taken from the cut specimens. The best agreement between in situ and the cut hardness data were those obtained from S2. However all the other microhardness measurements did fall in a very narrow band (Fig 64, 65).

S2 had hardness values in the range of 600 - 650 Hv. This corresponds well with the observed microstructures for S2 which contained large temper carbides and acted to toughen and harden the steel. Given the coarser carbide structure observed for the blade material of S2 when compared to the other blade materials from this group, the higher hardness observed may be taken as an indication that S2 has a higher carbon content. The wrought iron back on S4 had an average hardness of 180 Hv.

3) Anomalies

As mentioned, S4 was anomalous because it did not have a brass backing. Instead it had a wrought iron back. This structure was similar to the other wrought iron pieces, such as in the plane blades and the mortising chisels. Also, as noted, the steel structure of S2 was somewhat different from the remainder of the group because it contained a more obvious dispersion of temper carbides.

B) Brass

1) LOM

The brass backings that were examined were from S1, S2, S3, S5 and S6. All metallography was performed in situ with the exception of S3 which as explained earlier had a very small piece cut from the parent metal in an area underneath the handle. The brasses examined were similar to each other and therefore allow extrapolation from the one piece which we were able to closely examine (the cut piece from S3) and made some reasonably reliable conclusions concerning the entire group.

The typical microstructure obtained for the brass backings consisted of large grains with fairly straight annealing twins typical of a cold worked and annealed structure. The mounted brass piece cut from S3 showed only these twins (Fig 66, 67) while the other brass backings which were examined in situ showed some slip traces indicative of light cold working.(Fig 68) It appears that the brass was first fabricated into grooved rod-stack and the saw was then tapped into the backing groove.

There were two main types of inclusions in the brass observed. The first type was spherical and was found inside the grains. The second and larger type was globular and was found either along the grain boundary edges or at triple points.(Fig 69) These two types of inclusions were identified through XRD and are discussed below in the chemical analysis section.

2) Chemical Analysis

Our goal in this section was to identify the chemical constituents of the two types of inclusions found in the brass backing materials. Both types of inclusions were very soft when compared to the surrounding matrix, so a special polishing technique had to be developed which would inhibit pullout. This procedure was discussed in the experimental procedure section. Microprobe analysis by Xray wavelength dispersive spectrometry (WDS) and energy dispersive spectrometry (EDS) found both of these inclusions to contain only elemental Pb.(Fig 70, 71) This agrees well with our optical observations which noted that both types of inclusions were slate gray in color and behaved similarly under polarized light. This finding enabled us to categorize these brasses as a leaded brass.

3) Hardness

The in-situ hardness values of the brass backings were as would be expected for an alpha brass of this type. They measured around 150 Hv and were comparable in hardness to the wrought iron backing used in S4. The brass and the wrought iron backings therefore performed similarly in that the hardness of the wrought iron was only about 30 Hv higher than the brass. Thus the two types of backing materials were essentially interchangeable, and both performed their task reliably and had similar mechanical properties.

IV)Socket and Tanged Chisels

The socket (SC) and tanged (TC) chisels all had a steel edge forge welded to the cutting surface of the tool. A specimen was taken from the steel edge of TC4 for further analysis. Two pieces were also cut from SC1, one from the edge which revealed the interfacial area between iron and steel and one cross section from the socket end which revealed much about the forming processes of these chisels. Like the mortising chisels and plane blades, the TCs and SCs had an area of tempered martensitic steel forge welded to a wrought iron or low carbon steel body. The expectation was therefore to see a diffusion interface for carbon with varying types of microstructures depending of the amount of carbon present. As will be shown below, difficulties encountered in the hardness data and in interpretation of replica structures prevented a clear determination of the interfacial structures.

A) Tanged

1)LOM

LOM revealed heavily scaled and corroded areas on all the tanged chisels. The steeled cutting edges were very similar to those previously seen in other tools. As described below, a feathery tempered martensitic structure with small single phase slag inclusions was the norm. These slag inclusions were elongated and finely dispersed throughout the observed areas. Again, as the interface was approached we saw larger laths of martensite which gradually changed into ferrite grains.(Fig 72) The slag on the wrought iron side of the interface was elongated, but too heavily corroded in any of our in situ observations to examine its structure or chemical composition. They closely if not exactly resembled in shape the two phase slag inclusions seen in other specimens. However, no cut pieces were taken from the wrought iron side of a tanged chisel. Luckily we were able to cut a small section from the steel edge of TC4.

Results from the steel edge of TC4 showed a fine grained tempered martensitic structure with one phase slag inclusions and retained austenite.(Fig 73) The slag inclusions were elongated and indicated forging in the direction of elongation. Running nearly the entire length of the specimen was one extremely large slag inclusion. This slag inclusion was also elongated in the same direction as the smaller ones found in the same matrix.(Fig 74) Xray spectra showed that the qualitative composition of the matrix and fold slags was the same except that the amount of Fe in the matrix slag was slightly higher than in the "fold" slag. Because the slag particles in the matrix were only about 2 μm in diameter, we concluded that the high Fe counts in the matrix slag were a result of the beam interacting with the iron matrix surrounding the slag particles. In this case, the interaction volume was larger than the size of the particle in question.(Fig 75, 76)

The remaining tanged chisels, TC1 and TC3 were also examined using LOM of replicated specimens. Some indication of the presence of an interfacial region was also seen for TC1 but this was not a very good replicated image. Although features suggestive of martensite in the steel, ferrite in the body and mixed structures at the interfacial region were noted, the general quality of the replicas was poor due to heavy corrosion. In only one case (TC3) was a clear image of martensite near the interfacial region.(Fig 77) TC3 also showed an iron/steel type of interface similar to that described above.(Fig 78)

2) Hardness

The hardness measurements made across the interfacial regions for both TC and SC groups also proved to be problematic. Our hardness measurements did not concur with optical findings for TC3 where a martensitic structure was observed.(Fig 79) In this specimen, and as well for TC1 and TC4, hardness values remained consistently low. These incongruent findings were probably due to the large amount of scale on the outer regions of this tool. When our in situ hardness tests were performed, we took measurements directly from the outer surface of the tool. No hardness measurements were taken from TC1 due to the slant of the bevel. Overall, the hardness of the tanged chisel group was around 200 - 300 Hv for the wrought iron bodies and up to 400 Hv for the steeled edges. The very low readings in the steeled edges of the tanged chisels were probably due to heavy scale and corrosion. The hardness values should have climbed to the level of that observed in the mortising chisels, namely around 700 - 800 Hv. This was indeed confirmed by performing additional hardness tests on a specimen cut from TC4. This piece came from the tip or steel edge and had a hardness around 700 Hv.

B) Socket Chisels

1)SC1

a)LOM

SC1 was a socket chisel which had two specimens cut from the parent piece, one from the socket end and one from the cutting edge. "SC1 tip" consisted of a piece of steel bonded to a piece of wrought iron. The steel was located along the cutting edge of the piece. The steel side of the specimen appeared to have a banded structure when etched in a 4% picral solution.(Fig 80) At high magnification both light and dark bands contained lath martensite.(Fig 81) There were small single phase elongated slag inclusions which are also in the direction parallel to the bands. This specimen was etched with Oberhoffer's etch to reveal changes in phosphorus concentration in the steel.(Fig 82) Oberhoffer's etch does not attack high phosphorus regions and the previously noted banding was therefore attributed to phosphorus segregation.(Fig 83) Oberhoffer's etch was able to delineate the lath martensite in both light and dark areas. This specimen was also etched with Stead's etch which also confirmed phosphorus segregation. Stead's etch does not attack phosphorus rich regions so the light bands were high in phosphorus while the dark bands were low in phosphorous. Stead's etch brought out the lath martensitic structure only in the low phosphorus regions.(Fig 84)

The wrought iron side of the interface contained ferrite grains with globular shaped single phase slag inclusions.(Fig 85) The rounded shape of the slag inclusions indicated that the piece was not extensively forged. Also in this ferritic wrought iron area was what appeared to be a ghost or remnant structure. Upon step etching in 2% nital and Marshall's etches, this ghost structure was more clearly brought out and was determined to be epitaxial or halo ferrite. Large pockets of lath martensite which contained globular two phase slag inclusions formed in the wrought iron body of the tool near the interface where carbon was able to diffuse from the steel side.

The specimen cut from the socket end of SC1 has a variable microstructure.(Fig 86) The center region consisted of pearlite (dark) with quite a bit of ferrite (light) and contained small spherical slag inclusions while the outer regions were ferritic and contained larger quantities of elongated slag inclusions.(Fig 91) The pearlitic region could be present due to local carburization at the surfaces of a hammer weld seam or could be due to a local variation within the wrought iron although this was not observed in the wrought iron bodies.

Phosphorus segregation was also noted at the socket end of the chisel within the ferrite regions. As before, the specimen was etched using picral, Stead's and Oberhoffer's etchants. It was noted that Oberhoffer's etch and Stead's etch both tended to etch low phosphorus regions (dark) without etching high phosphorus areas (light).(Fig 87, 88) Picral tended the reverse the contrast effect or give a "negative image" of the same areas.(Fig 89, 90) Because all of the etching techniques give only an indirect indication of the presence of phosphorus it was decided to compare these results with direct determination of phosphorus by microchemical analysis.

b) Chemical Analysis

In order to directly confirm the presence of phosphorus in the specimen cut from the socket end of SC1 we used electron probe micro analysis (EPMA). WDS Xray maps for phosphorus gave a positive identification for the segregation structure. (Fig 92, 93) These maps also show a very high phosphorus content to be present in the two phase slag and some of the small spherical slag particles.

SC1-tip and SC1-socket were subjected to further hardness testing in order to correlate microhardness measurements with regions of high and low phosphorus content. A hardness profile was obtained for SC1-tip which began in the ferritic body and moved into the banded steel tip was obtained. (Fig 94) Assuming that the lighter etching bands were high in phosphorus we anticipated a higher hardness readings from these areas, and this was confirmed. Average hardness was about 650 Hv in the high phosphorus areas and about 600 Hv in the low phosphorus areas. SC1-socket which also showed segregation of phosphorus within its microstructure produced hardness readings of about 250 Hv in the high Phosphorus ferrite areas and around 225 Hv for the low phosphorus ferrite areas.(Fig 95)

2) SC2

a) LOM

The only other socket chisel examined was SC2 (SC3 was dropped from our analysis because it was heavily corroded). SC2 was in situ polished and etched for examination by LOM and was then replicated. The interfacial area of SC2 was very similar to that observed in SC1, with the exception of the region just on the steel side of SC2. In this region the single phase slag stringers were seen which were elongated in a direction parallel to the interface. (Fig 96) The wrought iron region of SC2 appeared to consist mostly of epitaxial ferrite.(Fig 97)

b) Hardness

Hardness tests performed on SC2 and SC1 also produced similar results.(Table 5)(Fig 79) The wrought iron side of the interface had a hardness of around 300 Hv for both specimens. The steel in SC2 was slightly harder (approx. 610 Hv) than the steel in SC1 (approx. 680 Hv). This was probably due to a slightly different carbon content or tempering condition for the two steels. Nevertheless, hardness increased further as the indentor moved across the interface.

Discussion

I) Chisels

A) Compare and Contrast Types

Three types of chisels were examined in this study. As described in previous sections the mortising, tanged and socket chisels consisted of a low carbon wrought iron body with a higher carbon steel cutting edge. During manufacture, the body was created first and the steel edge was then forge welded in place. After welding, the piece was put through a series of heating and quenching steps.(Fig 7) A mortising chisel is designed to cut slots for mortise and tenon joints. The resulting slot is the same size as the width of the blade. Hence, these types of chisels were made in a variety of shapes and sizes.(Fig 6) The tanged chisel derived its name from the shape of the back end of its blade. These tanged chisels were general purpose chisels that were used for chopping and paring. The socket chisel were constructed in response to the demand for a heavier duty chisel that could be used by carpenters, millwrights and wheelwrites. (GAY '93). The back end of a socket chisel was forged into a socket shape so the handle could be inserted to achieve a sturdier overall construction.

Mortise chisels generally had a long, thin blade that had uniform width over the entire length of the chisel. Socket and tanged chisels usually had a wide edge and became more narrow at the handle. Because the socket chisel was a heavier duty chisel, it's overall appearance was somewhat larger in all dimensions than the tanged chisel. Microstructurally, all three types of chisels contained similar edge and body materials. They all have a harder, stronger cutting edge which was forge welded to a tougher, less expensive body. Our hardness data show a slightly higher overall hardness for the steel edges of the mortising chisels as compared to the steel edges of the socket and tanged chisels.

B) Correlate Historical to Metallographic Observations

1) Manufacture

The chisels were all made in a fashion similar to that outlined in the processing and manufacturing section of this thesis. First, the body was formed and then the cutting edge was forge welded to the body. After forming the chisel, the entire piece was subjected to a series of heating and quenching steps. Our microstructural observations confirm the previously described manufacturing process. In the body we saw elongated two phase slag stringers which visually confirm the forging that the pieces underwent during forming. This area consisted of ferrite grains and in some cases, subgrain boundaries even developed. The presence of these subgrain boundaries pointed to some type of plastic deformation during hot working. This agreed well with what we know about the extensive forging that these pieces underwent. In the body of some of the specimens, especially the socket chisels, we also noted the presence of epitaxial ferrite in some of our specimens. This is due to the relatively slow cooling that took place when the ferrite grains formed. The metal was heated into the ferrite and austenite phase region and allowed to cool slowly. During this process the ferrite which precipitated from the austenite formed on the undissolved ferrite. (SHR '66) Also of interest was a segregation structure that was seen in a piece cut from the socket end of SC1. The unusual variation in microstructure seen in the micrographs was inferred to be caused by impurity segregation during solidification. Stead's and Oberhoffer's etch were used to bring out this segregation structure which first appeared as a ghost structure with the more traditional 4% picral etch. Xray microprobe analysis at a number of different point locations confirmed the segregating element in question to be phosphorus. An Xray map was taken of a pertinent area and again confirmed the presence of phosphorus. Because the segregation structure cannot be seen without a proper etch, LOM of the mapped area were taken for a more exact correlation.(Fig 92, 98) As can be seen in Figure 87, Stead's etch revealed the high Phosphorus areas in our map to appear lighter and the surrounding low phosphorus matrix to appear dark.

The presence of an interdiffusion region between the steel edge and the wrought iron body could be seen as a white hazy line in many micrographs.(Fig 39) As carbon diffuses from the higher carbon edge to the lower carbon body a gradation in microstructure developed and was observed in all the chisels. Because the steel edge was forge welded to the body there was no barrier to inhibit the diffusion of carbon. During subsequent hardening and tempering these interdiffusion areas of medium carbon content developed a large lath martensite structure. This will be discussed further in the next section.

In general, the slag inclusions found in the steel edges were consistently different in nature than those slag inclusions found in the body. The body typically contained a two phase slag which consisted of a partially crystallized silicate matrix with dendrites of FeO. The presence of excessive quantities of FeO in the slag would indicate that the furnace conditions were not sufficiently reducing, hence the oxidation of iron in the slag generated a wustite second phase rich in iron. Of interest is the observation that the FeO dendrites were often found to initiate their growth at the interface between the slag and the surrounding metal.(Fig 99, 100) This is because the higher Fe content of the metal when compared to the slag created a greater potential for the oxidation of Fe. The dendrites then continue to grow into the silicate slag. These two phase slags are also much longer and more elongated than the slag seen in the steel. Both the slag in the steel and the slag in the iron contained a plethora of elements.(Fig 34, 35) The major qualitative difference between the two is the higher phosphorus content found in the slag present in the iron. This would lead one to believe that these two materials (steel edge and iron body) were smelted from two different ores or that processing has had a major effect on the slag chemistry. Studies made on single objects out of context such as those examined here cannot however be used to infer such specific information.

The body of MC8 contained several elongated and broken slag inclusions.(Fig 101) The ends of the broken slag inclusions are rounded but the surfaces of fracture within the inclusion are straight and jagged, which is indicative of a brittle fracture mechanism. By measuring the length of the slag stringer before deformation and after deformation (length of slag stringer plus the spacings between breaks) it was possible to estimate the percent reduction in area the last time the piece was deformed using the simple relationship,

$$\Delta l / l_o = [.0432\text{mm} - .035\text{ mm} / .035\text{ mm}] * 100\% = 24.42\ \%$$

where,

Δl = the change in length of slag particle due to deformation ($l_f - l_o$)

l_o = length of slag particle before deformation

l_f = length of slag particle after deformation.

The steels seen in the chisels all appeared very similar. Although the steel edges were considerably more free of slag than the iron portions of the chisels, the obvious presence of slag inclusions within the steel suggests that this material was probably a blister steel. Although crucible (cast) steel was first discovered and used in 1835 it did not find its way into common use until 1875.(BAR '81b)(Fig 102) These steels were homogenous in their microstructures but in some cases they were found not to be homogenous in chemical composition. After etching with Stead's etch it was noted that some of the steels contained layers with differing amounts of phosphorus. This can be seen in Figure 19 by observing the alternating dark and light bands. A correlation was found to exist between phosphorus content and hardness.(Fig 94) In the lighter bands where phosphorus content was high, the hardness was higher. A simple t-test (KUR '91) was applied to the resulting data to confirm that there is a real statistical difference between the two sets of data (high and low phosphorus). The banded structure associated with phosphorus segregation together with the presence of slag inclusions provided strong support for the identification of these steel edges as fabricated from blister steel. This banded structure was observed in every steel edge cut from a chisel (SC1, MC1, MC2, MC7, MC12 and MC13) except TC4.

Further hardness tests were performed on SC1-socket which contained phosphorus segregation in a lower carbon steel.(Fig 95) Again we observed a higher hardness in ferritic areas with higher Phosphorus. These hardness measurement differences were also confirmed by a t-test to have a statistical difference between ferrite areas of high and low phosphorus. The steel edges in some of the chisels contained retained austenite which can be seen as white specs in both SEM and LOM for TC4.(Fig 73, 103)

The slag inclusions in SC1-socket and the single phase slag in SC1-tip were both analyzed microchemically. The SC1-socket sample was taken from the socket end of SC1 while the SC1-tip sample was taken from the steel edge of SC1. SC1-socket contained both single phase and two phase slag inclusions, both of which are high in phosphorus. The single phase slag present in SC1-tip had no measurable phosphorus present. This is understandable when considering the fact that the body of SC1 was a low quality inexpensive low carbon wrought steel which only served the purpose of providing a tough body for the tool. The only area which need be of good quality was the cutting edge, and in this case the lack of phosphorus found in the steel is indicative of the use of a higher quality steel. In a similar fashion, the piece taken from the edge of TC4 contained a homogenous steel structure which was of an even better quality than the edge of SC1. The slag in the steel edge of TC4 contained no phosphorus as in the other cut chisel pieces and also did not display the banded structure cause by the segregation of phosphorus when etched with Oberhoffer's etch as was seen in SC1-tip. The higher quality of TC4 was also reflected by the higher hardness readings (about 50 Hv higher) obtained for TC4 in comparison to SC1-tip.

Despite the common occurrence of Apatite (calcium phosphate and other Phosphorus minerals) in ore mineral deposits worldwide and the propensity of P_2O_5 to segregate strongly to SiO_2 (MAX '78) and its deleterious effect on steel processing, very little quantitative work appears in the literature. Mihok et. al. noted varying hardnesses in early roman swords that contained the phosphorus segregation identified by etching with Oberhoffer's etch.(MIH '93) They reported a hardness of about 180 Hv in high phosphorus areas and about 140 Hv in areas of low phosphorus. Gordon reported lower hardnesses (avg. 147 Hv) in ferritic areas of wrought irons vs. higher hardnesses (avg. 249 Hv) in areas of higher phosphorus.(GOR '92)

As mentioned earlier the slags in the wrought irons examined in this study were consistently very high in phosphorus. From the SiO_2/P_2O_5 phase diagram it can be seen that P_2O_5 tends to segregate to the uncrystallized portion of the slag.(LEV '64) (Fig 104) These would be seen as dark in the matrix region of the slag. However, due to the small size of the dark or uncrystallized areas, which were beyond the microchemical resolution limit for EPMA, we were not able to obtain good Xray spectra to confirm this. Gordon noted similar slag structures in his paper entitled "The quality. of Wrought iron evaluated by EPMA". He saw a wustite second phase that precipitated out in a matrix of partially crystallized fayalite. His results show a higher content of P_2O_5 in the dark or glassy (uncrystallized) areas of the slag than in the fayalitic matrix.(GOR '84).

2)hardening and tempering

As shown in the processing and manufacturing section (Fig 7) toolmakers in eighteenth century Europe and in Colonial America already had a well developed processing technology by which chisels were put through a series of heating and cooling procedures during their manufacture. The first heating step after forge welding was a long one day anneal. This promoted the diffusion of carbon from the steel edge into the body. We confirmed this by observation of the microstructural variation in the interfacial region between the steel edge and the iron body as seen in all of the chisels. Just on the low carbon side of the diffusion interface (the original weld seam) we observed the presence of pockets (prior austenite grains) which contained large lath martensite. This martensite probably formed during a second heating and the subsequent water quench from about 1500 ° F which was performed to harden the steel. After hardening the steel, it had to be toughened by another heating process called tempering. This was reflected in the microstructure of all the chisels examined. The steel edges of these chisels are composed of a tempered martensite. Tempered martensite stains during etching due to the carbide present while untempered martensite is more difficult to etch and remains bright. Some tempered structures were more feathery than others which can be explained by slight differences in carbon content. For example, the results of a carbon analysis performed by Laboratory Testing Inc. in Dublin, PA were done by combustion analysis and showed the steel edge of TC4 to contain 1.12% carbon. This carbon level would produce hardness values that are well within our hardness values reported from TC4 which were around 700 - 750 Hv. From a graph of hardness vs. tempering time we were able to confirm the results of the combustion analysis. (ANO '78) A comparable steel quenched from 1600°F and tempered at 480°F, which is close to the 1500°F austenitizing temperature and the 518°F tempering temperature suggested to be in common use in the eighteenth century would have a range of 650 Hv for the shortest tempering time (as quenched) to 940 Hv for the longest tempering time (100 hours) reported. TC4 therefore can be approximated to have been tempered for around .33 hours or about 20 minutes, consistent with our understanding of good practice.

The hardness tests performed on the socket chisels were somewhat inconclusive because all hardness tests were initially taken in-situ, the large amount of pitting corrosion and scale on the outer areas of the socket chisels had a significant effect on our results.(Fig 79) We believe that the corrosion caused the softer body to yield a higher hardness and the harder edge to yield a lower hardness. Although we did see an increase in hardness as the indentor moved from the body into the edge, we relied on microstructural information and the cut specimens for our conclusions. An attempt was made to grind more material from the polished areas of the socket and tanged chisels to reveal the metal underneath but the pitting type corrosion was so deep that no new results were obtained. Our basis for comparison here was the hardness profile obtained from one of the cut specimens. This result, obtained from SC1-tip, is reliable and corresponds well with the data from the other types of chisels.

The hardness tests performed on the tanged chisels were even more inconclusive than those from the socket chisels. Looking only at the hardness results (Fig 79), one would assume that these chisels contained no steel edge. This is obviously not true when the data from the TC4-cut specimen and the microstructures are considered. All the tanged chisels examined contained an interfacial region with remnant, highly corroded martensitic structures near the interface. The body contained ferrite grains and some epitaxial ferrite with elongated two phase slag stringers, as did the socket and mortise chisels.

As explained in the experimental procedure section on all of the mortising chisels in-situ hardness tests were performed along the length of the top side of the blade. These hardness readings included areas of low hardness in the chisel body all the way through the interfacial region and into the hardened steel edges. The result was a series of profiles which plotted hardness as a function of distance which can be seen in Figures 21-23. A "typical" example of a hardness profile for one of the mortise chisels is MC1.(Fig 21) The presence of tempered martensite in the steel edge caused hardness to increase to between 700 and 800 Hv. The interfacial region immediately next to the steel edge consisted of large lath martensite and caused the hardness profile to drop until hardness readings level off again at around 175 Hv in the ferritic body. MC1, MC2, MC3, MC6, MC7, MC11 and MC12 all had similar if not identical hardness profiles and accompanying microstructures.

Anomalous hardness behavior was observed in MC4, MC5, MC8, MC9, MC10 and MC13. MC4 had a steel edge with a hardness of only about 425 Hv. This is explained by noting that the microstructure in the steel edge is not martensite but pearlite with a considerable amount of ferrite.(Fig 41) Also at the very edge of this chisel, hardness readings dropped off to that of the ferritic body (around 150 Hv). This is probably a result of repeated sharpening of the chisel which caused the very edge to wear off leaving only the ferritic structure of the iron which was beneath the steel. MC5, MC8, MC9 and MC10 all had higher hardness in the bodies than in the bodies of the other chisels. This was because these chisels did not have a purely ferritic wrought iron body as did the others but rather had a body constructed using a higher carbon pearlitic iron. Also the appearance of the interfacial region in these chisels was different than in the other mortising chisels. The interfacial region of these chisels could not be photographed because the change in structure from high carbon to low carbon occurred over such a large distance.

The tip of MC10 had a lower hardness than would be expected for a martensitic steel. The microstructure of the steel edge of MC10 reveals a more of a feathery structure and which would indicate a lower carbon content, a slower quench rate or even a longer tempering time. MC5 also had a lower hardness steel edge, approximately 645 Hv. However, microstructurally the martensitic steel edge of MC5 is similar microstructurally to the other steel edges with higher hardnesses. We can only postulate that the heat treatment for these two chisels (MC10 and MC5) was in some way different than the majority of the group, although the steel and iron portions indicate that they are potentially equivalent tools to the remainder of the group.

II) Molding Plane Blades

A) Correlate Historical to Metallographic Observations

1) Manufacture

All the planes examined in this study are of the molding type and were used at one time to cut decorative shapes in wood molding. Molding plane blades appear to be simple tools but our analysis showed a more complicated construction. As with the chisels, all the plane blades had a steel cutting edge that was forge welded to a lower carbon steel body.

The planes contained a low carbon ferritic body with a higher carbon martensitic steel edge. The body consisted of ferrite with two phase slag stringers which were elongated during forging. The interfacial region between the steel edge and the lower carbon body contained a gradation in microstructure depending on the relative amount of carbon present. Of interest was the absence of martensitic laths at the interface as seen in most of the chisels. PB3, PB5 and PB6 all had a thin oxide layer at the interface. This oxide was probably entrapped there during forging. Hence, these pieces contained a barrier for carbon diffusion at the interface in the form of an oxide. The steel edge in these plane blades was a tempered blister steel with only small slag inclusions. The one sacrificial piece, MPS3, did not show banding and it did not appear to be properly heat treated and quenched. We did not investigate this piece further, i.e.., to determine if it was blister or cast steel, but the latter seems improbable. Typically, all the plane blades appeared to have steel edges free of phosphorus and wrought iron sections which are high in phosphorus in the slag inclusions. However, the single phase slag in the steel edge of MPS3 consisted mainly of an aluminosilicate.(Fig 60)

2) Hardening and Tempering

The hardness of the steel tips of the plane blades were approximately 600 Hv. This was about 200 Hv lower than the hardness values measured for the chisel group. The lower hardness of the edges of the plane blades can be understood by examination of the interfacial region of these pieces. The plane blades which were examined and were without an oxide layer at the interface contained no martensitic laths, opposite to the case for the interfacial regions of most of the mortising chisels where martensite was observed within the interfacial region. The reason for this absence of martensite at the interfaces probably was related to their carbon content and to the temperature at which the pieces were annealed. For example, the steel edge of MPS3 (the sacrificial plane blade) contained martensite with islands of ferrite which showed epitaxial regrowth. After the original anneal at approximately 1500°F, the entire piece was probably heated into the two phase region instead of being fully austenitized at a higher temperature before the quench. The carbon content of the steel was high enough to form martensite in the edge but not high enough to form martensite in the interfacial region. Due to the high Ms of the steel pieces, the smith was able to obtain some martensite when he quenched from the two phase region. (SAM '80) Based on the presence of ferrite in the steel edge of MPS3 and referring to the phase diagram (Fig 57) the composition of the steel in this piece has to be than the eutectoid composition, .8 weight percent carbon. By examination of the iron carbon phase diagram it can be seen that if the austenitizing temperature was 1500°F (815°C), which is the temperature indicated (Fig 7) as the most probable austenitizing temperature the carbon content would have to be .3 weight percent or less in order to obtain the martensite plus ferrite structure..

To obtain a better estimate of the % carbon in the steel edge of MPS3, we austenitized the steel at 950°C for 20 minutes then slow cooled down to room temperature. The resulting structure was pearlite and ferrite. The % pearlite was obtained by quantitative computer image analysis which calculated the area fraction of the pearlitic areas seen in red.(Fig 59) Ten trials showed the % pearlite to be 48%. Thus, from the Fe-C phase diagram, we used the lever rule to estimate the carbon content of the steel to be around .4 weight percent. Hence the smith must have been below an austenitizing temperature of about 815°C to obtain martensite plus ferrite from the quench. Also, from our estimation of % carbon (.4 %) and our hardness result it appeared that the hardness values obtained were consistent with a tempering temperature somewhat below the recommended 518°F tempering temperature. The presence of epitaxial ferrite within the ferrite islands in the steel again confirmed an austenitizing temperature of 815°C.

All of the plane blades exhibited similar interfacial regions and similar hardnesses in both the steel edge and the iron body. We also believe that these blades were annealed for a considerable time after welding, because of the gradual change in carbon content with distance across the interface. The variation in hardness in the wrought iron bodies of the plane blades was probably due to the high and variable amount of phosphorus present. The two plane blades with higher hardnesses in their bodies (PB1 and PB5) are well within the hardness ranges reported by Gordon for wrought irons containing phosphorus.(GOR '92) He points out that high phosphorus areas in ferritic wrought iron can cause the hardness to increase from 147 Hv to 249 Hv; our data from the ferritic iron bodies of PB1 and PB5 are around 200 Hv. Although this piece was not etched with Oberhoffer's etch a ghost structure could be seen when it was etched with a 4% picral solution consistent with the presence of epitaxial ferrite and phosphorus segregation.

Also of interest is the consistency within the plane blade group. The steel edges of all the plane blades fell in the 600 Hv range while the hardnesses of the iron bodies fell in the 150 - 200 Hv range. Also the interfacial regions in all the plane blades were remarkably similar.

Hardness profiles were obtained for PB6 and PB2 which display and compare the behavior in the body and the edge of both tools. PB6 was a piece with oxide at the interface and its hardness profile changed from very hard to very soft over a distance of only a few millimeters. PB2, which did not have oxide present at the interface and therefore did have a good metallurgical bond at the interface, had a more gradual decrease in hardness with distance. When comparing the hardness profile from PB2 (which had no oxide at the interface) to the hardness profiles for the mortising chisels we noticed that the plane blade profile had a much more gradual decrease in hardness with distance than do the mortise chisels profiles. There was no immediately obvious explanation for this consistent difference. It could be that the difference in annealing time was only related to a difference in established practice, or could be because the heftier tool was more susceptible to heavier use and more frequent failure.

III) Saws

A) Compare and Contrast Types

The saws examined in this study consisted of six backsaws, both dovetail and sash. We were able to examine sections of the blade material and a small area of the backing material for each saw. As described earlier, the backing material for all the saws was brass, with the exception of S4 which had an iron backing. All the saws in this group, except one, had similar microstructures and microhardness with the exception of S2. The blade material of S2 had a much higher hardness and a distinctly different microstructure of large temper carbides present in the martensite. This will be described in more detail in the section on hardening and tempering. For now it should suffice to say that S1 and S3-S6 are constructed out of similar steel materials. All five of the brass backings examined are virtually identical both in microstructure and in hardness results. The iron backing also had a hardness that was approximately equal to the brass backs and consisted of a ferritic structure with a small amount of pearlite. Thus, from a technical standpoint both backing materials (brass and iron) could be used interchangeably. Possibly the low carbon wrought iron backing was chosen by the smith for financial reasons.

B) Correlate Historical to Metallographic Observations

1) Manufacture

In 18th cent. England, there were two main methods for the production of saw blades. The steel was either rolled out and then case carburized for a number of days (blister steel) or produced from a crucible steel bar. (TYL '80) With the level of examination utilized in this project, it was not definitely possible to infer which of the two steel types were used when fabricating the saws, but the presence of elongated slag particles favored a blister steel. Also, the presence of elongated single phase slag suggests some type of forging or rolling in the direction of elongation. The slag consisted of a single phase calcium iron aluminosilicate along with other trace elements. This slag was similar to that seen in the steel edges of the chisels and plane blades. All microstructures observed in the blade material of the saws were that of a tempered martensite. The saws were thus rolled, annealed, quenched and tempered, and finally ground and or polished for finishing.

As mentioned above, the backing for S1, S2, S3, S5 and S6 was made of brass. The microstructure for these backings were all similar and contained large grains with slightly bent annealing twins.(Fig 66) This indicates that there was only slight additional deformation or performed after the final anneal, possibly associated with tapping the steel blade into the brass grooved backing. We cannot tell how many times the pieces were deformed and annealed, just that the brass piece was in the annealed condition just prior to assembly into the saw. The in-situ pictures taken from the saw blade materials showed slip lines due to localized deformation while the cross sectional view of the brass piece cut from the backing of S3 did not show these slip lines. In all probability this was also due to the fact that the brass backing was hammered or clamped into place, and thus introduced slip lines into the microstructure of the outer surface and left the inner areas relatively unaffected. There was no high Zn β phase present in the microstructure, so we could safely assume that the composition is 29% Zn or less. (Fig 8) This corresponded well with what we know about 18th cent. brassmaking. One of the most common brasses produced during this time was called a yellow brass and had a composition of 70/30. (TYL '80) Yellow brass later became known as cartridge brass and was one of the softest brasses made and was mainly used for deep drawing and spinning.

In the 17th and 18th centuries, the brass industry had two main ores available for its use. The more plentiful of the two, ZnS, was very difficult to smelt and its use had to wait until a technology developed that could cope with the high S content of zinc sulfide. The less commonly found but more frequently used ore was Smithsonite or $ZnCO_3$. (TYL '80) This ore was easier to smelt because it did not contain the deleterious sulfur present in ZnS. By understanding the use of these ores we were able to hypothesize that $ZnCO_3$ must have been used to produce the brass backings found on the saws. Qualitative chemical analysis of inclusions present in the brass backing did not show any residual S anywhere. The large round inclusions and the smaller ones on the grain boundaries proved to be elemental lead.(Fig 70) If ZnS were used as an ore, PbS inclusions or even residual ZnS might be expected to have been observed instead. Finally, of interest would be an approximation of the % cold work that these brasses underwent. By considering the hardness and assuming that these brasses were initially in the annealed condition, we estimated that 20% reduction or less was incurred, consistent with the microstructural observation. (ANO '72)

2) Hardening and Tempering

As mentioned previously, the homogenous martensitic microstructure of the saw blade materials together with only a very small slag inclusion content indicate the use of a very clean high quality steel. Hardness tests did show the blade of S2 to have the highest hardness. The first tests were performed in situ along the edge of the blade; these hardness readings were later confirmed on a cut specimen. (Fig 64) After observing the microstructure of S2 and noting the presence of large temper carbides (precipitated cementite) which the other blades did not have, we concluded that S2 was a steel with a higher carbon content when compared to the others in the group.

A small specimen was also cut from the blade of S3. The in situ hardness readings and the hardness data from the cut piece of S3 correlated well. They both were about 450 Hv, which was not as high as S2. Again, the only difference in microstructure between S2 and

S3 was the presence of large temper carbides in S2, probably resulting from a higher carbon content.

A piece was also cut from the blade of S6. The hardness data from the cut piece of S6 fell in the range 350 - 400 Hv but the in situ measurements were as high as 500 Hv. Similar problems were encountered in the socket and tanged chisel groups and was interpreted to be caused by the presence of scale or pitting type corrosion on the outer surface of the piece. This was also believed to be the explanation for the results of S6.

Conclusions

Stead's etch and Oberhoffer's etch both give a consistant and equivalent indication of the presence of variable phosphorus in many of the wrought iron and steel materials examined during the present study. The presence of phosphorus in one wrought iron specimen (SC1) was confirmed directly by Xray mapping during electron probe microanalysis. The increase in hardness of ferrite due to the presence of phosphorus was also directly confirmed. Detailed quantitative analysis of slag inclusions were not performed during this study. It is recommended that in the furture, the detailed microchemistry of the slag inclusions be measured by microprobe analysis and compared with other specemens known to be made from cast steel or blister steel. Similarly, phosphorus segregation should be studied by Xray mapping in the banded steel structures.

Saws

The saws examined in this study were very consistent in terms of microstructure and hardness with the exception of the Manwaring saw (S2). The saw blade material is in all cases studied here probably a blister steel. Their microstructures are homogenous with almost no visible slag inclusions. Hardness measurements were around 600 Hv for all the steel sheets, and hardness measurements varied no more than 80 Hv from one area of a blade to another. The material used for the blades of the saws are thin sheets of steel which were probably rolled, annealed, quenched, tempered and finished from a larger bar or ingot.

1) The Manwaring saw from London (S2) had superior hardness explainable by the presence of temper carbides in the microstructure and probably related to a higher carbon content.

2) In order of decreasing hardness the saw identities are as follows : S2 (London), S5 (London ?), S3 (Sheffield), S4 (Sheffield), S1 (Birmingham) and S6 (Sheffield).

3) The Spear saw from Sheffield (S6) which is identified as "cast steel" has a uniform martensitic structure and a low slag inclusion content as do all the other saws. Therefore the marking is not inconsistent with the structure seen, but the very heavily worked condition of all the saws prevents a definitive comparison.

Backing Materials

brass

The brass backings examined all appear to be made form yellow leaded brass which has a composition near Cu 70% - Zn 30%. The microstructure shows annealing twins with Pb inclusions at the grain boundaries and at triple points. The in-situ micrographs show slip lines indicative of light working which were probably introduced when the backing material was hammered into place. Based on the hardness of the brass backings, it was estimated that the material experienced 20 % cold work or less. The hardness measurements for the brass backings were very consistent for all the saws examined.

low carbon iron

The backing material on the Kenyon saw (S4) is a low carbon wrought iron. Hardness measurements indicate that its hardness is comparable with the brasses used in the other saws; the hardness of this low carbon wrought iron is only about 50 Hv higher than the brasses.

Plane Blades

The plane blades examined in this study contained a wrought iron body with a steel edge that was forge welded in place. The steel edges are all fully martensitic while the wrought iron bodies are all fully ferritic. The interfacial region between the steel edge and the iron body contains a graded variation in microstructure. Hardness measurements taken from the steel edges and the iron bodies of PB1 - PB6 show good consistency. There were no significant differences in fabrication or hardening/tempering procedures within this group. The only major observable difference within this group is the presence of a thin oxide layer at the interface between the steel edge and the iron body in PB3, PB5 and PB6. Normally this would indicate a poor weld but in this case the presence of an oxide at the weld seam hasn't affected the performance of the plane blades. Hence we conclude that this design performed adequately for the user; no major modifications were made over the time period examined.

Mortising Chisels

The mortising chisel group contained 14 chisels, all made by forge welding a steel edge onto a lower carbon wrought iron body. In most cases the steel edge is fully martensitic, there is a graded structure in the interfacial region and the body is fully ferritic. The martensitic steel tips are indicative of the quench hardening step for these tools. In each of the steel edges examined by cutting a small section from the bevel, a banded structure was noted in the steel edge and was shown to be due to the segregation of phosphorus. Thus the steel materials are in all probability a blister steel. In some cases (MC5, MC8, MC9 and MC10) the body of the chisel consisted of a higher carbon pearlitic iron. In these chisels, the weld seam was very clean which made it hard to see, even with a microscope. Several mortising chisels contain epitaxial ferrite in the tool body indicating a slow rate of cooling.

1) MC1, MC2 from Birmingham and MC3 from Sheffield are typical examples from the mortising chisel group. The presence of martensite in the interfacial region of these tools is evidence of a long time anneal after forge welding and prior to quench hardening.

2) The Newbould chisel (MC4) contained a ferrite plus pearlite edge with a ferritic body possibly indicating improper heat treatment sometime after the fabrication of the tools. The latest Newbould chisel (MC5) contained a martensitic steel edge with a pearlitic body and is therefore unusual in terms of the higher carbon iron used for its body.

3) The range of structures observed in the Newbould chisels (from Sheffield) examined could reflect changes in Sheffield products during the late eighteenth/early nineteenth centuries. The earliest Newbould chisel (MC3) has a ferritic body with a martensitic edge while MC4 has a ferritic body while the latest Newbould chisel (MC5) contains a pearlitic body with a martensitic edge. the pearlitic body would provide higher strength while not giving up toughness.

4) The Green chisels are very consistent, with the exception of MC7 which has a lower carbon ferritic body than the other Green chisels, containing a pearlitic body. The steel edges of the Green chisels have a fully martensitic structure. The steel edge of MC10 has a hardness that is about 200 Hv lower than the other Green chisels. This is either due to a lower amount of carbon or an excessive tempering time for MC10 in comparison with the other chisels.

5) The group of Law chisels (MC6, MC11) and the group of Green chisels (MC7, MC8, MC9, MC10) are more consistent in microstructure and hardness than the group of Newbould chisels (MC3, MC4, MC5).

6) The Law chisel (MC11) is very similar both microstructurally and in measured hardness to the earlier Birmingham examples (MC1 and MC2).

7) The wrought iron body of the James.Cam chisel (MC12) has an abundance of elongated slag inclusions, which indicate extensive forging normal to the direction of elongation. This chisel is similar to the other chisels examined in that it contains a steel edge forge welded to a wrought iron body. The presence of sub grain boundaries in the ferritic wrought iron body of this chisel indicates plastic deformation at hot working temperatures consistent with the known forging operation.

8) The James.Cam chisel (MC13) is marked "cast steel". The body of MC13 is a low carbon wrought iron and the steel edge is a uniform tempered martensitic structure. The label cast steel could only be referring to the steel edge. However, slag inclusions observed in the steel point to poor quality. More detailed analysis of the microstructue and study of slag inclusions is warranted.

Tanged and Socket Chisels

All socket and tanged chisels examined have a martensitic steel edge forge welded to a lower carbon wrought iron body. These tools were therefore forge welded, annealed and then quench hardened as wee the other chisels. The only not consistent with the rest of the group is TC4 which does not have a banded steel structure.

1) A comparison of the Birmingham and Sheffield examples is not entirely possible due to the deleterious effects of corrosion. However, SC1 (Birmingham) and SC2 (Sheffield) were examined and can serve as examples for this comparison. These two socket chisels

are similar in that they contain a steel edge forge welded to a lower carbon wrought iron body. Although SC1 was subjected to a much more detailed examination than SC2 there were no apparent differences between the two chisels, suggesting similar technology at both locations.

2) The "CS" stamp on the side of TC4 (Sheffield) could only refer to the steel edge not the wrought iron body of the tool. The uniform microstructure in the steel edge of TC4 combined with its high hardness suggests it to be of a higher quality steel than the other socket and tanged chisels and that it could be a cast steel. But this is very weak evidence considering the large amount of prior working which tends to homogenize cast structures.

Table 1

TOOL TYPE	LABEL	DATE	MARKS	ORIGIN
Dovetail Saw	S1	1767-1776?	Dalaway	Birmingham, England
Dovetail Saw	S2	1760-1810	J:Manwaring	London (?)
Carcass Saw	S3	1787-1821	Kenyon/Spring/London	Sheffield
Sash Saw	S4	1787-1821	Kenyon	Sheffield
Dovetail Saw	S5	early 19th century	Morman London/German	London (?)
Dovetail Saw	S6	ca 1816-1823	Spear	Sheffield
Molding plane blade	PB1	1730-1752	None	England
Molding plane blade	PB2	1750-1765	None	England
Molding plane blade	PB3	1748-1775	None	England
Molding plane blade	PB4	1766-1811	None	England
Molding plane blade	PB5	ca 1810-1830	Hildick	Sheffield
Molding plane blade	PB6	Before 1785	None	Massachusetts
Mortising chisel	MC1	Mid-18th century	Allen	Birmingham
Mortising chisel	MC2	Mid-18th century	Robert Moore	Birmingham
Mortising chisel	MC3	1770 (?)	Newbould	Sheffield
Mortising chisel	MC4	1770-1810 (?)	Newbould	Sheffield
Mortising chisel	MC5	1800-1825 (?)	Newbould	Sheffield
Mortising chisel	MC6	1770 (?)	P. Law	Sheffield
Mortising chisel	MC7	ca 1800	John Green	Sheffield
Mortising chisel	MC8	ca 1800	John Green	Sheffield
Mortising chisel	MC9	ca 1800	John Green	Sheffield
Mortising chisel	MC10	ca 1800	John Green	Sheffield
Mortising chisel	MC11	ca 1820 (?)	P. Law	Sheffield
Mortising chisel	MC12	mid 19th century	James.Cam	Sheffield
Mortising chisel	MC13	19th century	James.Cam/Cast.Steel	Sheffield
Mortising chisel	MC14	late 18th-early 19th	Weldon	Sheffield
Tanged chisel	TC1	1750-1770	Robert Moore	Birmingham
Tanged chisel	TC2	1770-1780 (?)	Gillot	Sheffield
Tanged chisel	TC3	1770 - 1800 (?)	P. Law	Sheffield
Tanged chisel	TC4	second half of 18th cent.	CS	Sheffield (?)
Socket chisel	SC1	1750-1770	Robert Moore	Birmingham
Socket chisel	SC2	1780-1820 (?)	John Green	Sheffield
Socket chisel	SC3	1770-1800 (?)	P. Law	Sheffield

Table 4

Table for Mortise Chisels

sample	H-avg steel	H-avg iron	origin	maker	structure-st	structure-ir
MC1	825	175	B-1750	ALLEN	M	F
MC2	750	175	B-1750	R. MOORE	M	F
MC3	800	200	S-1770	NEWBOULD	M	F
MC4	425	175	S-1770	NEWBOULD	P + alot of F	F
MC5	640	325	S-1800	NEWBOULD	M	P + F
MC6	725	175	S-1770	P. LAW	M	F
MC7	775	175	S-1800	J. GREEN	M	F
MC8	800	275	S-1800	J. GREEN	M	PS (fine sp.)
MC9	800	225	S-1800	J. GREEN	M	P + F@GB
MC10	600	300	S-1800	J. GREEN	FM	P + F@GB
MC11	800	175	S-1820	P. LAW	M	F
MC12	800	200	S-1850	JAMES.CAM	M	F with SGB
MC13	700	225	S-19 cent	JAMES.CAM	M with PAGB	F

Legend

FM = feathery martensity
M = martensite
P = pearlite
F = ferrite
PAGB = prior austenite grain boundary
GB = grain boundary
SGB = sub grain boundary
note : all hardness measurements have the units Hv

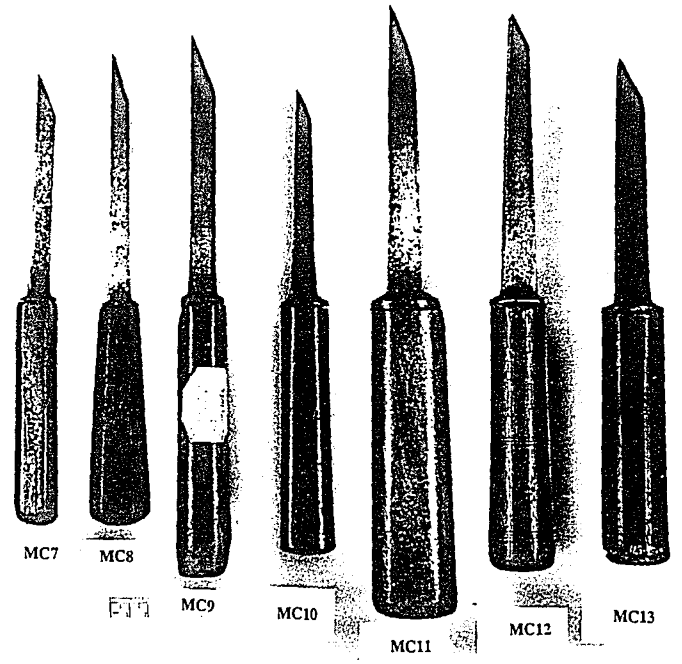


Figure 3

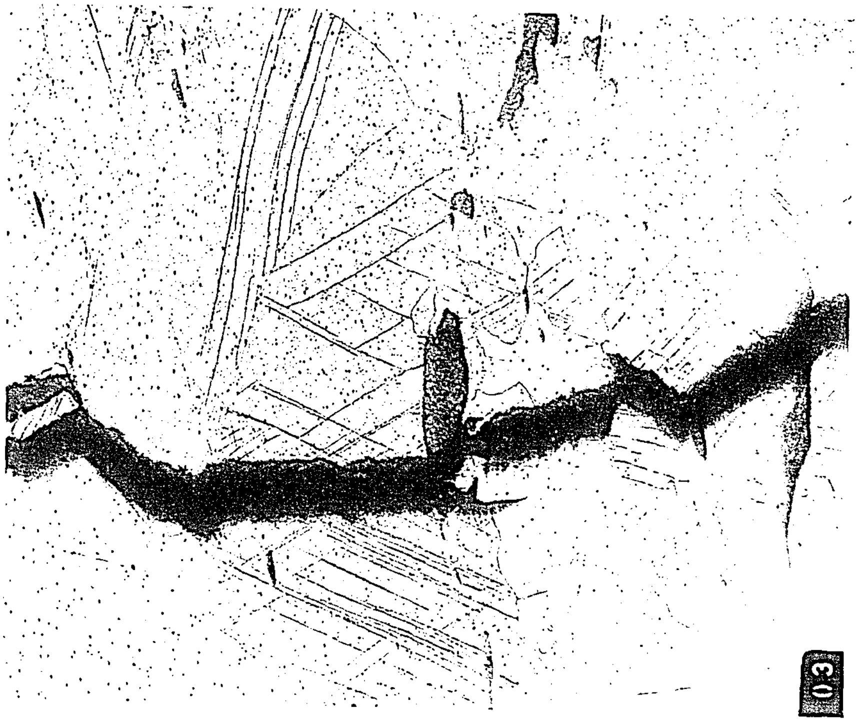


Figure 9 MC14, 50X, 4%picral

Figure 6 MC14, macroetch

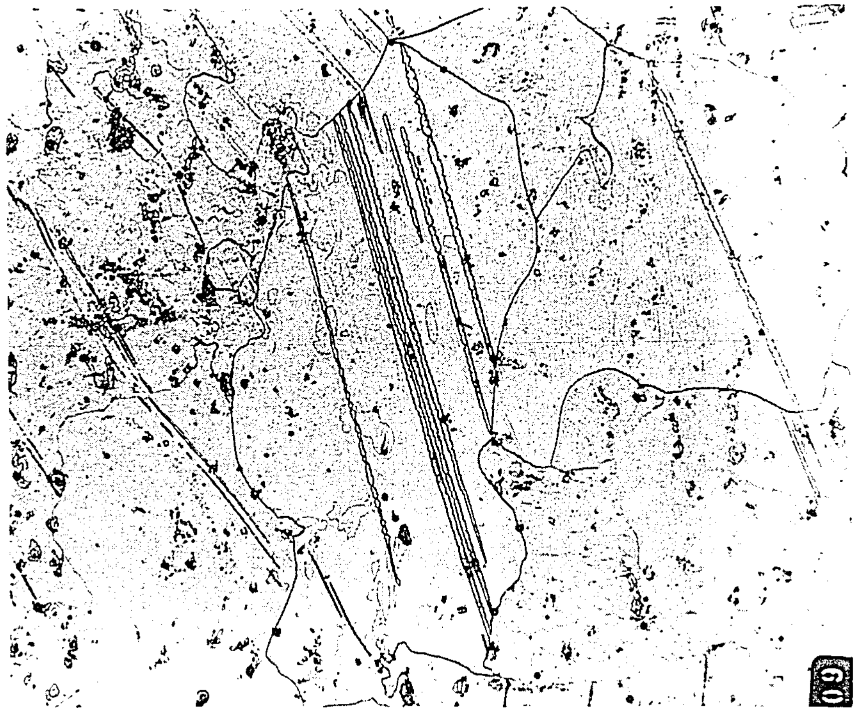


Figure 10 MC14, 200X, 4%picral

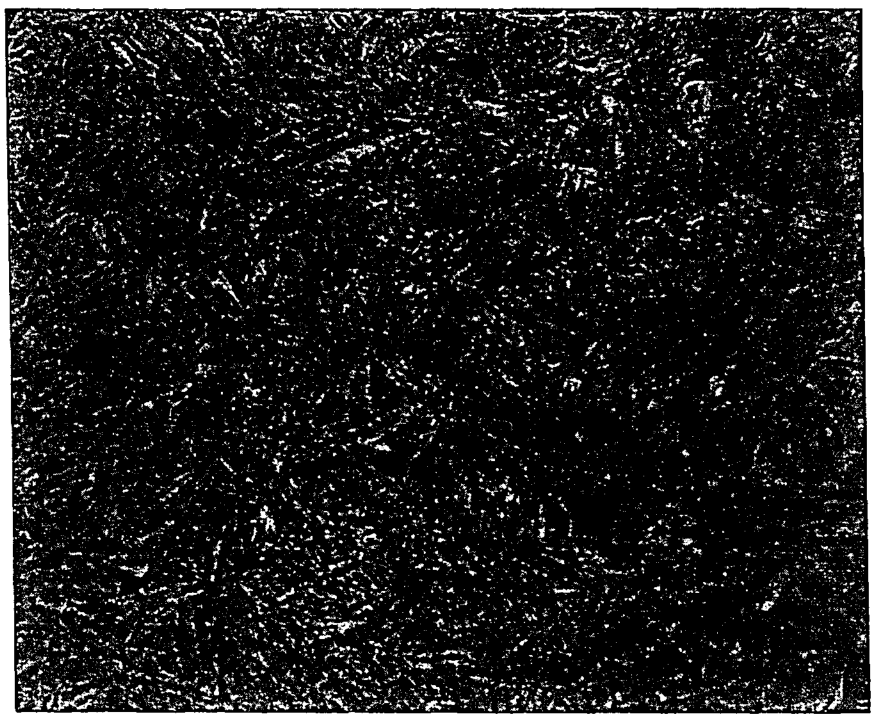


Figure 11 MC14, 1600X, 2%nital

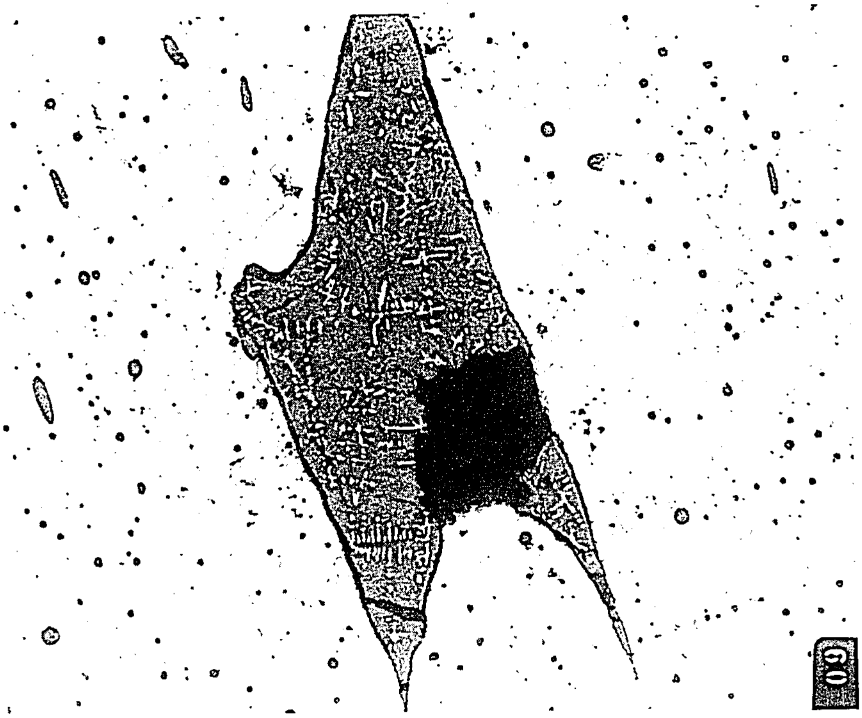


Figure 12 MC14, 200X, 4%picral

Figure 13 MC 12, 800X, Marshall's and 2%nital

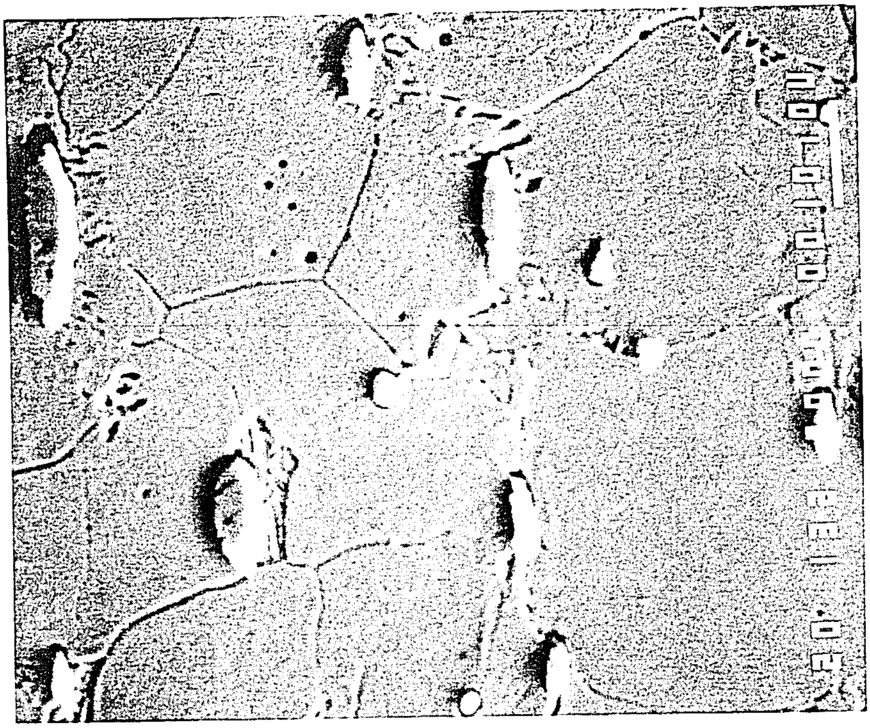


Figure 14 MC12, 2%nital, SE image

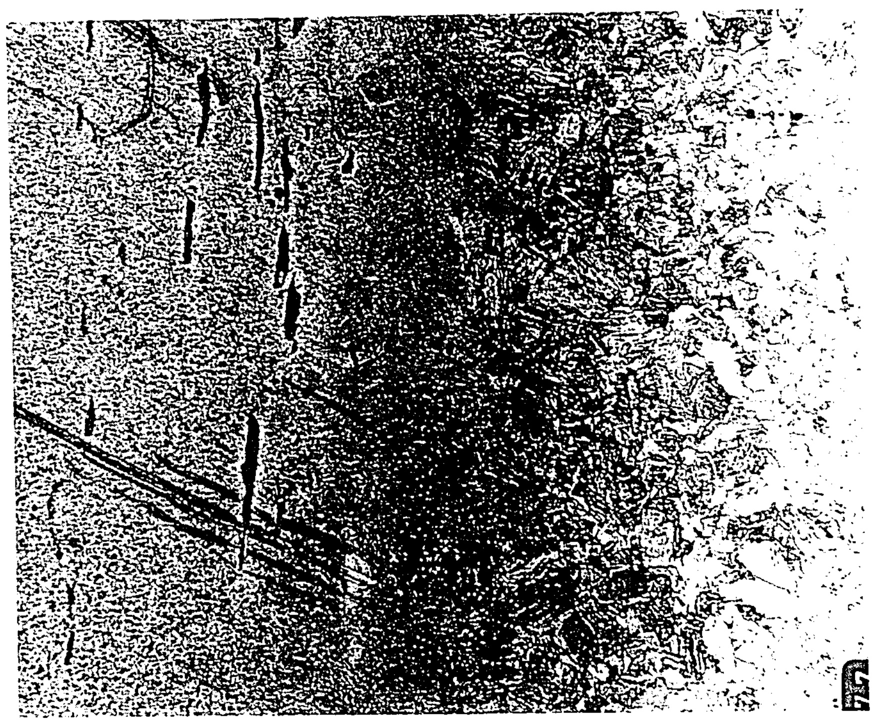


Figure 16 MC14, 160X, 2%nital

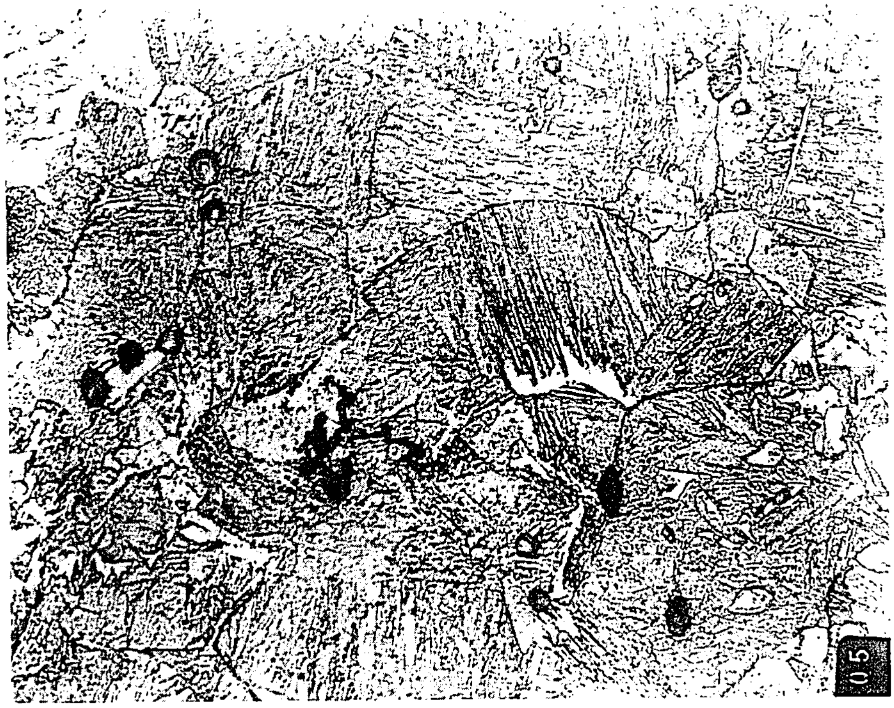


Figure 17 MC2, 1000X, Marshall's and 2%nital

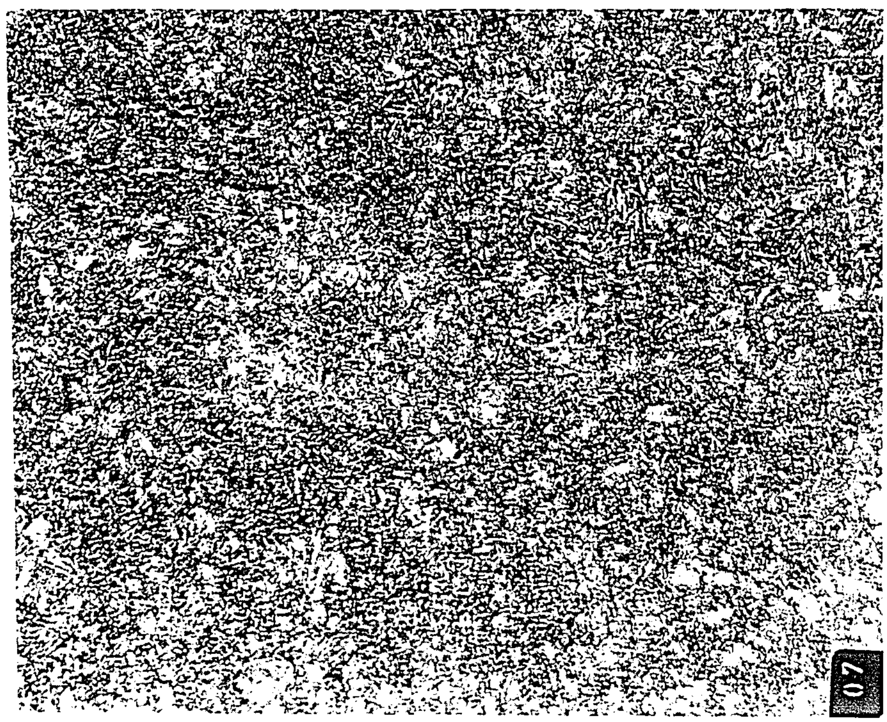


Figure 18 MC7, 375X, Marshall's and 2%nital

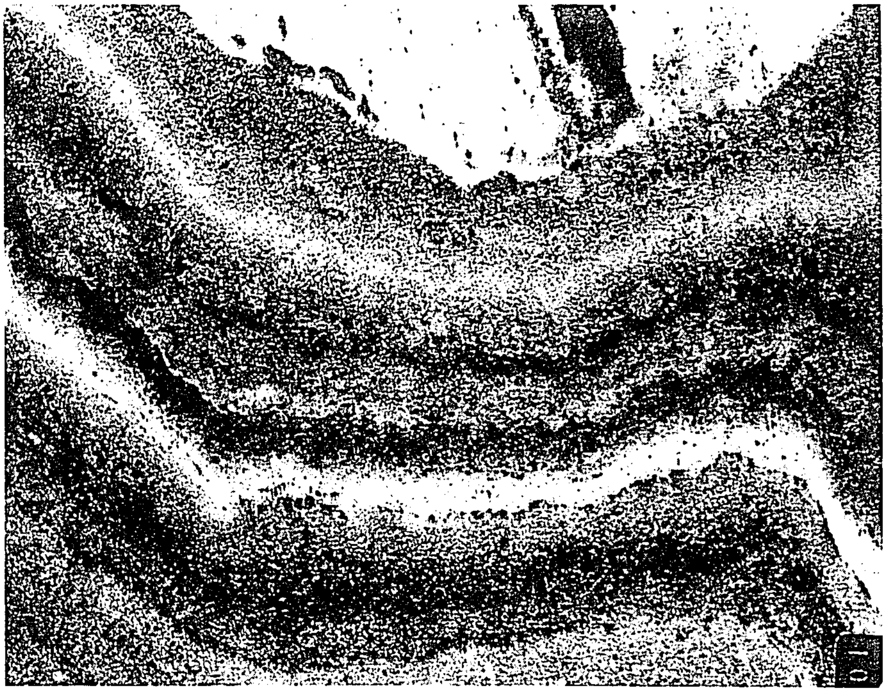


Figure 19 MC2, 50X, Stead's

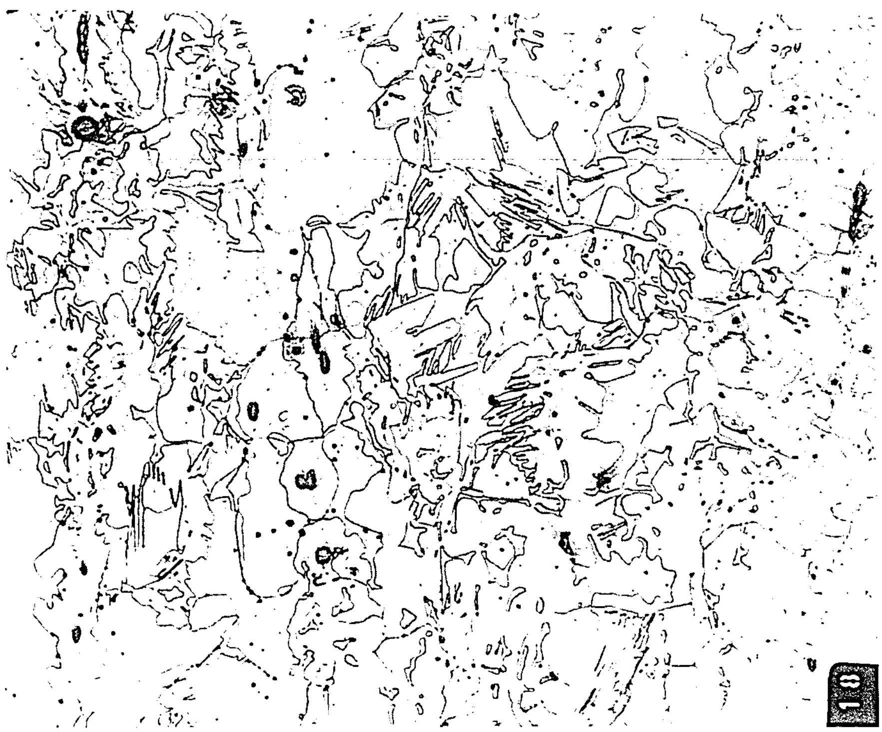


Figure 20 MC2, 250X, Marshall's and 2%nital

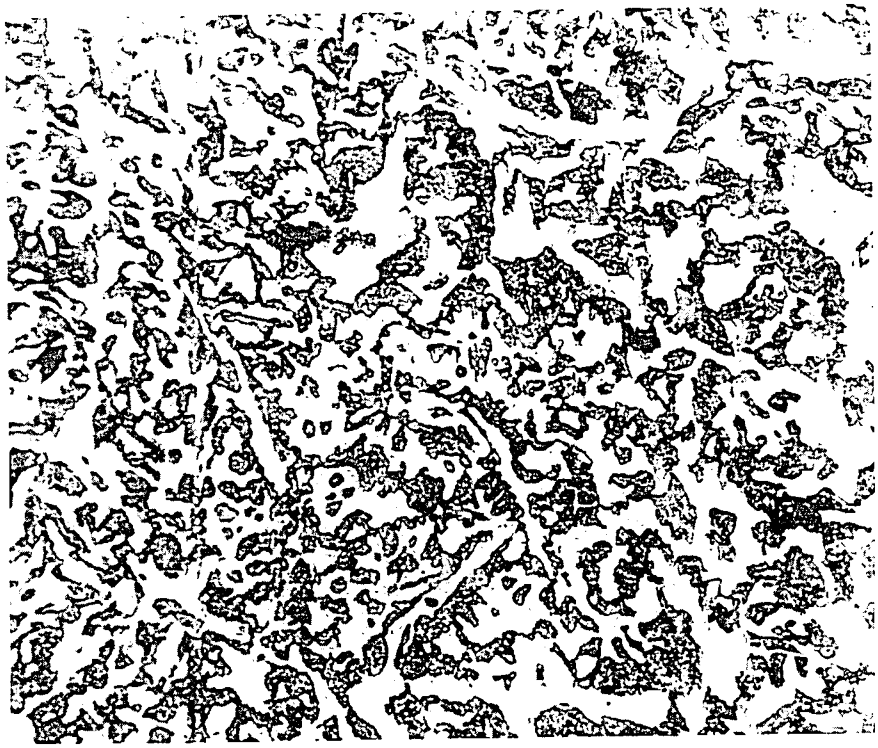


Figure 37 MC9 intermediate area, 600X, 4%picral

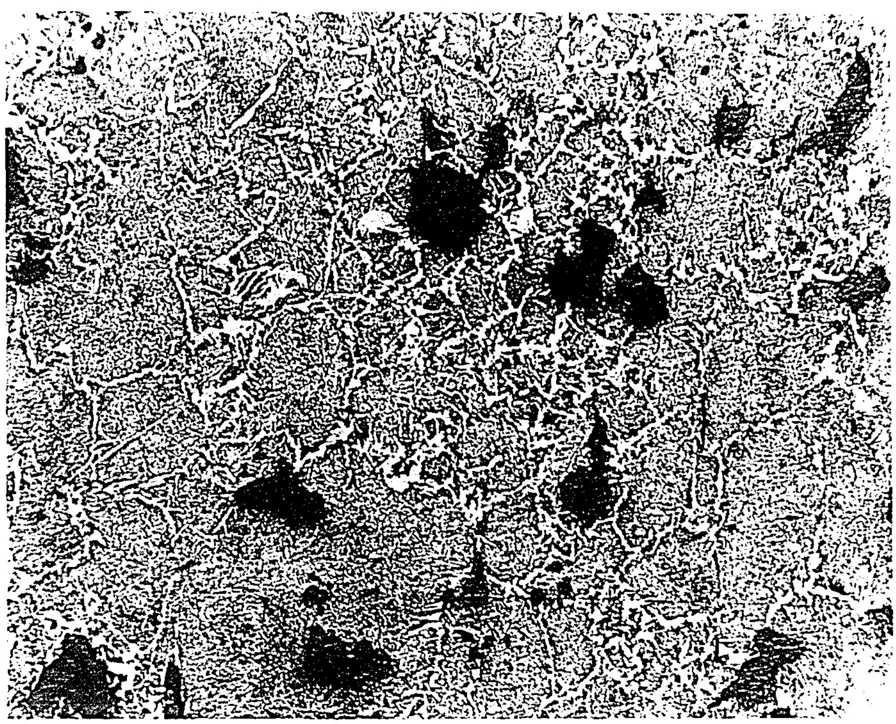


Figure 38 MC9, 150X, 4%picral

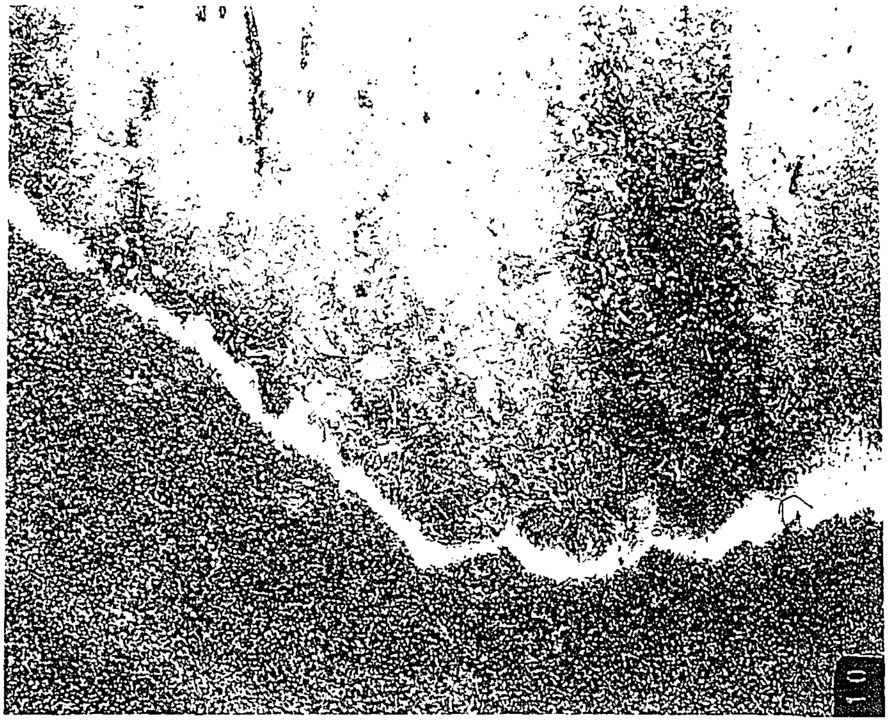


Figure 39 MC2, 50X, 2%nital

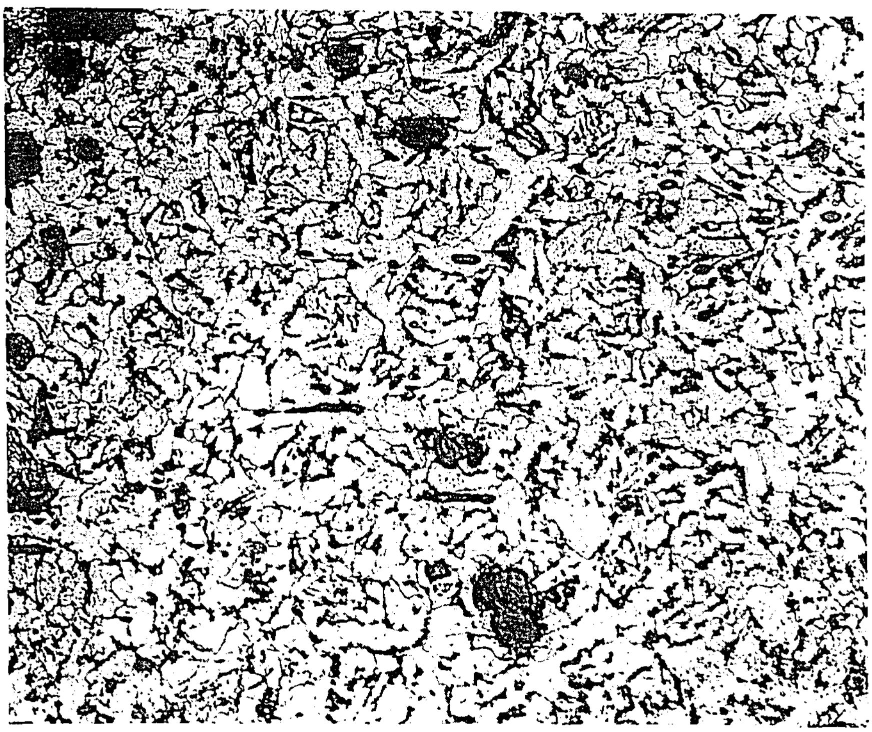


Figure 41 MC4, 300X, 4%picral

Figure 40 MC2, 500X, Marshall's and 2%nital

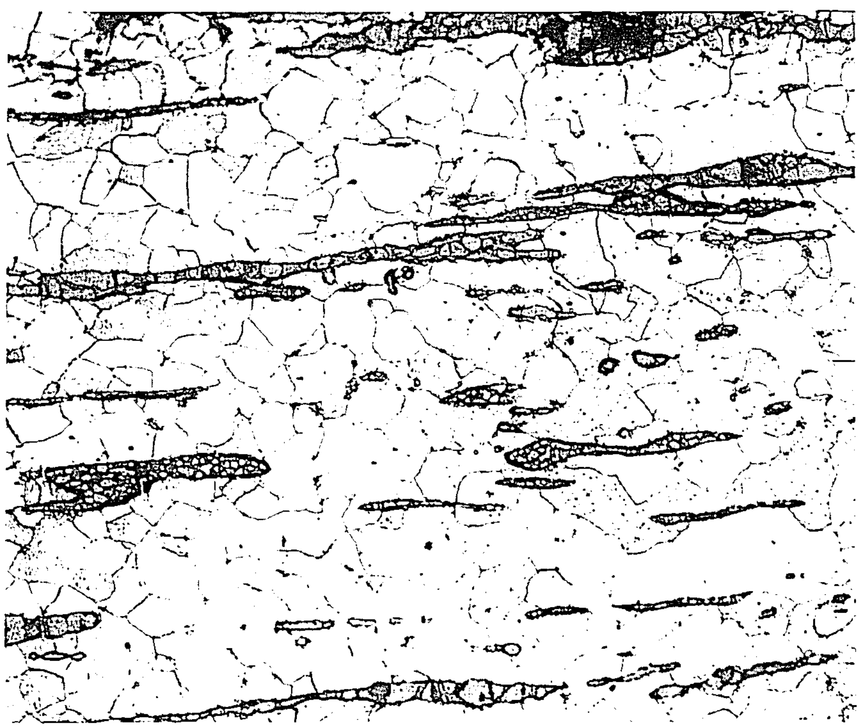


Figure 42 MC4, 300X, 4%picral

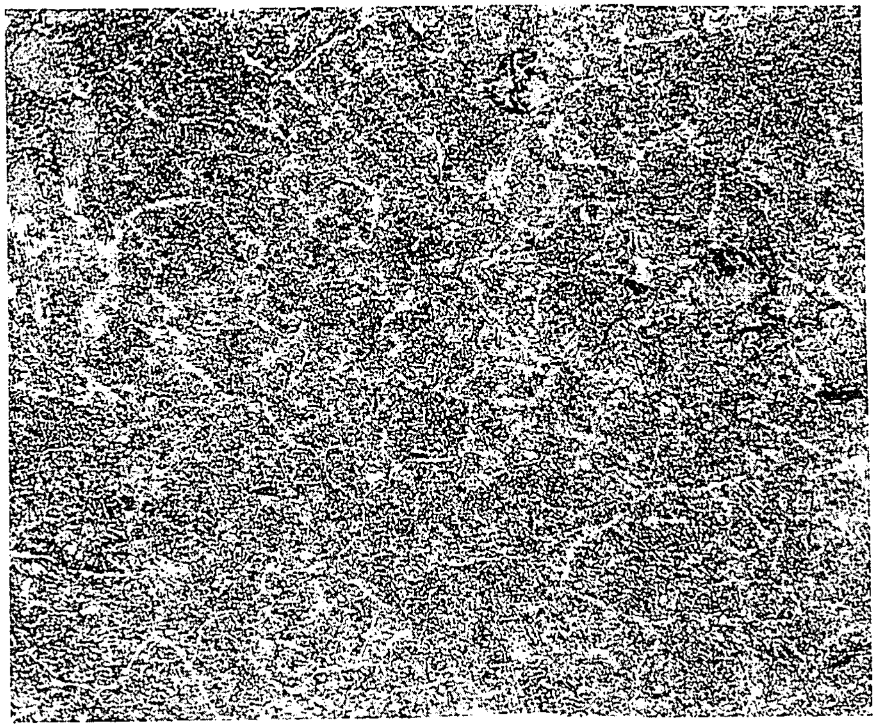


Figure 44 MC5, 300X, 4%picral

Figure 45 MC5, 800X, 4%picral

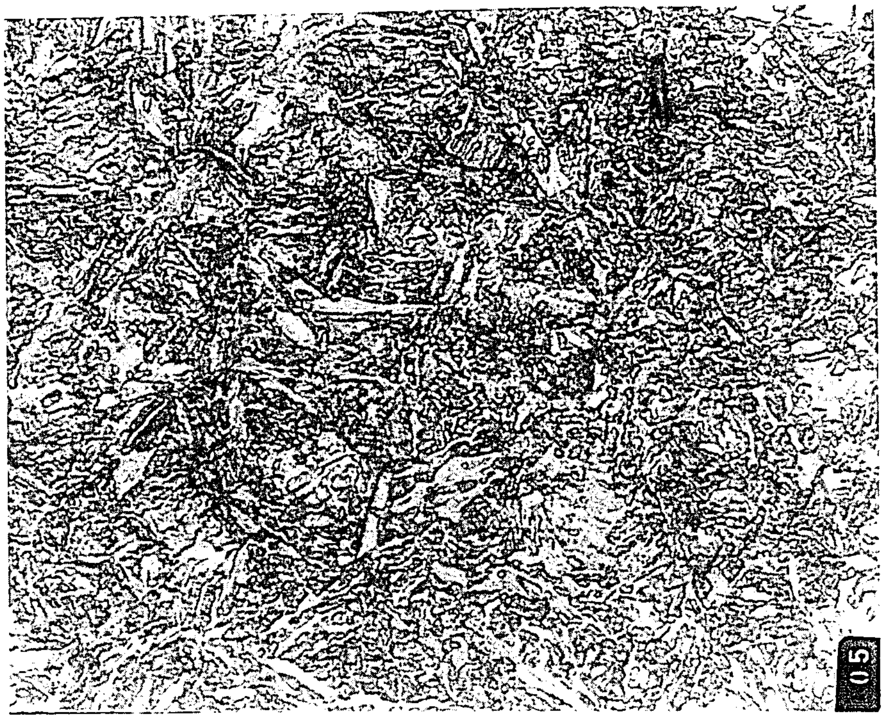


Figure 46 MC7, 800X, 2%nital

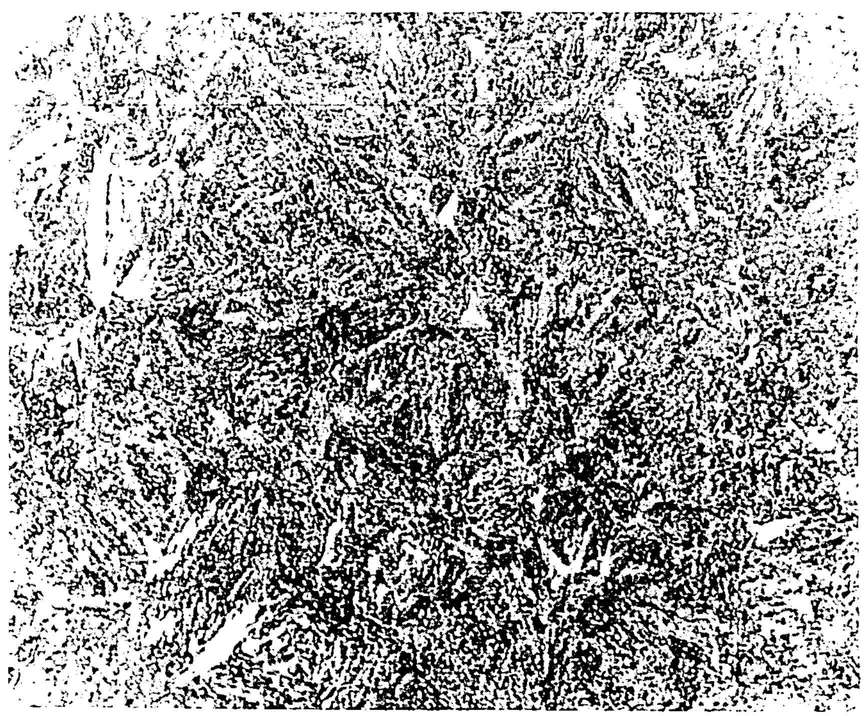


Figure 47 MC10, 800X, 2%nital

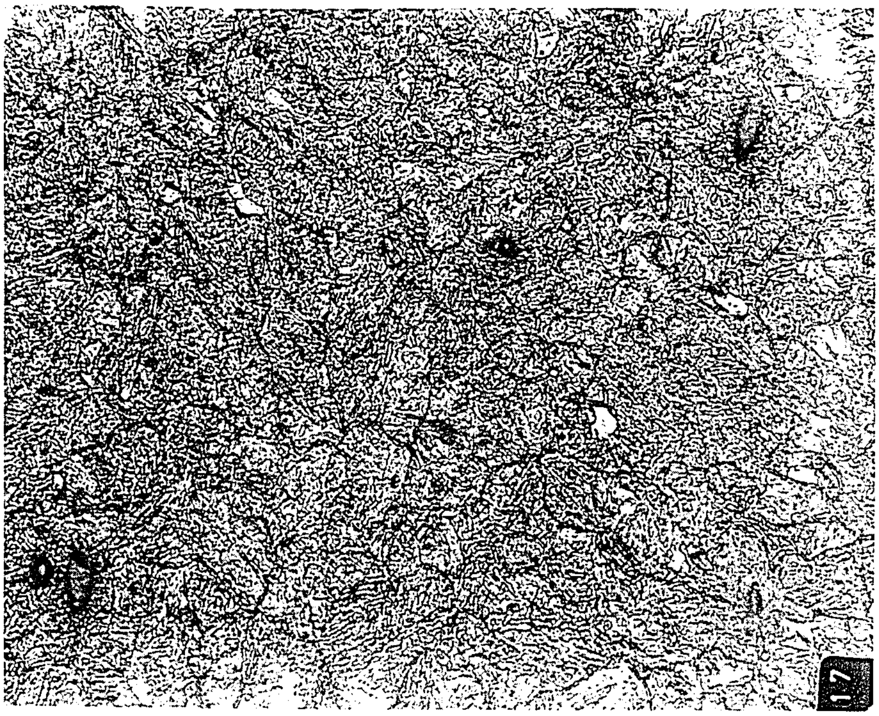


Figure 48 MC13, 850X, 4%picral

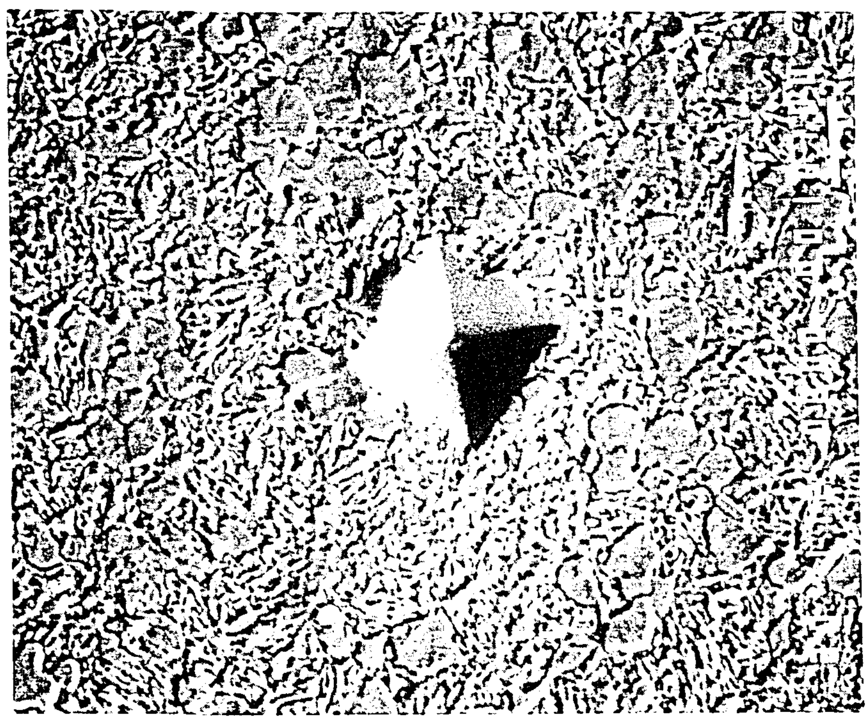


Figure 49 MC12, 4%picral, SE image

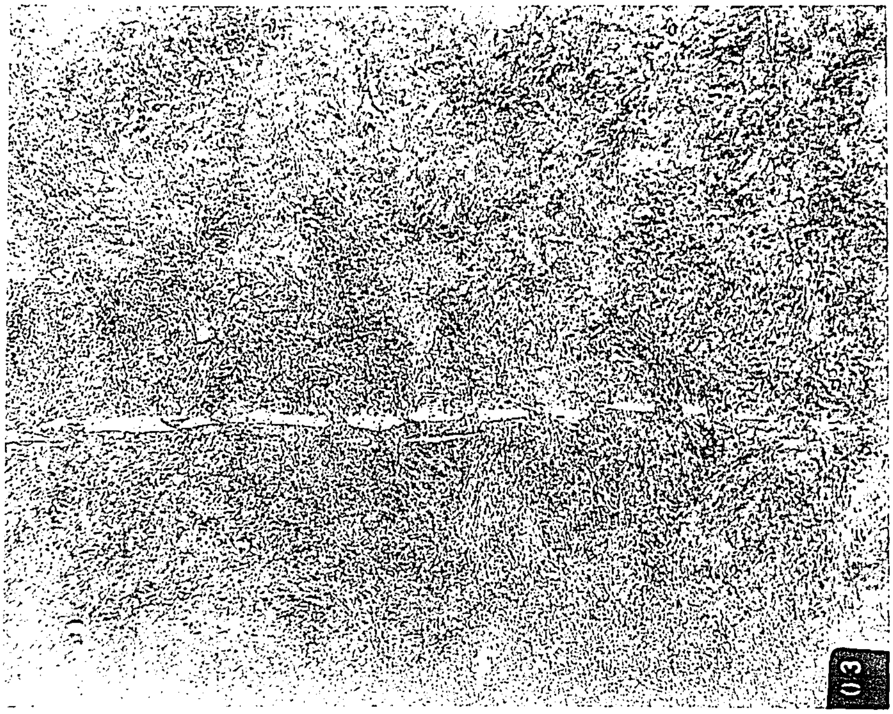


Figure 101 MC8, 250X, 4%picral

MC5

Figure 43 MC5

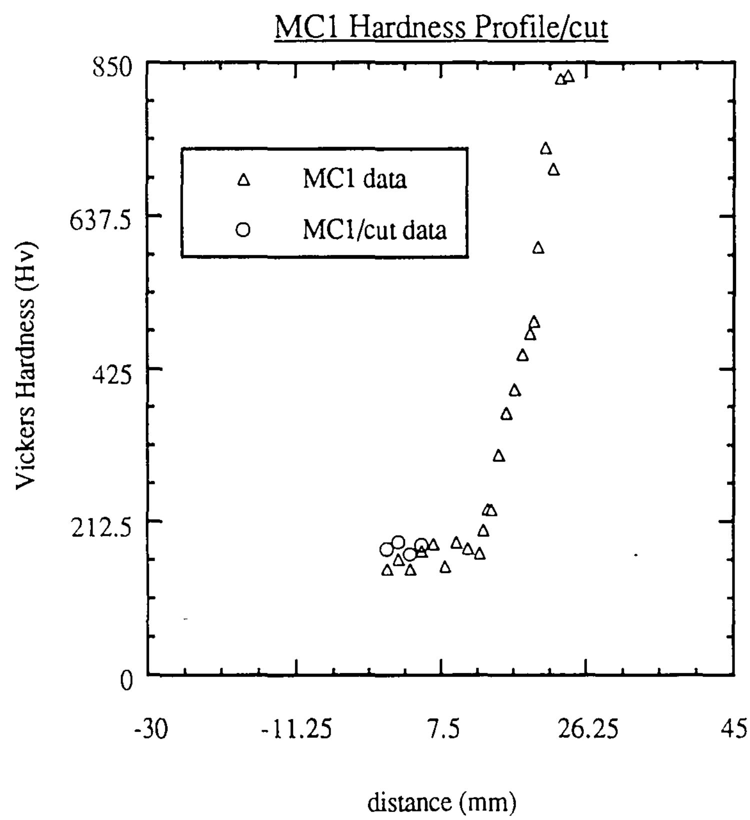


Figure 21 Hardness Profile

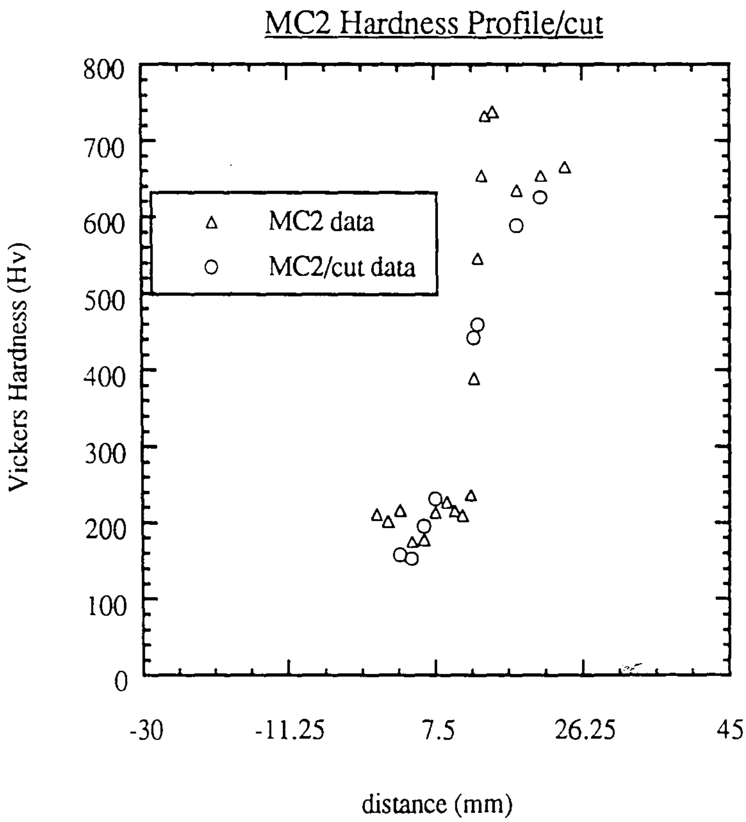


Figure 22 Hardness Profile

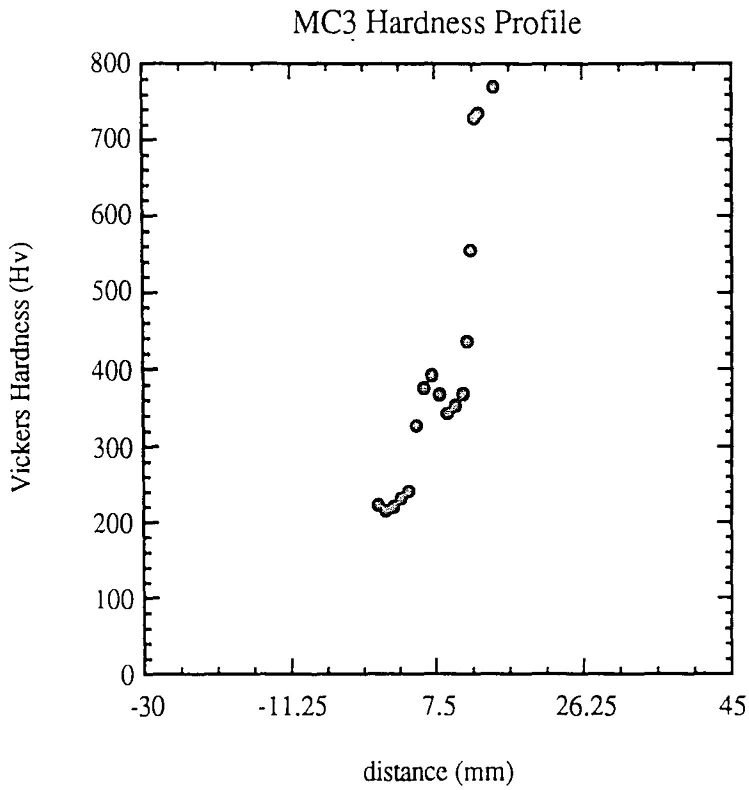


Figure 23 Hardness Profile

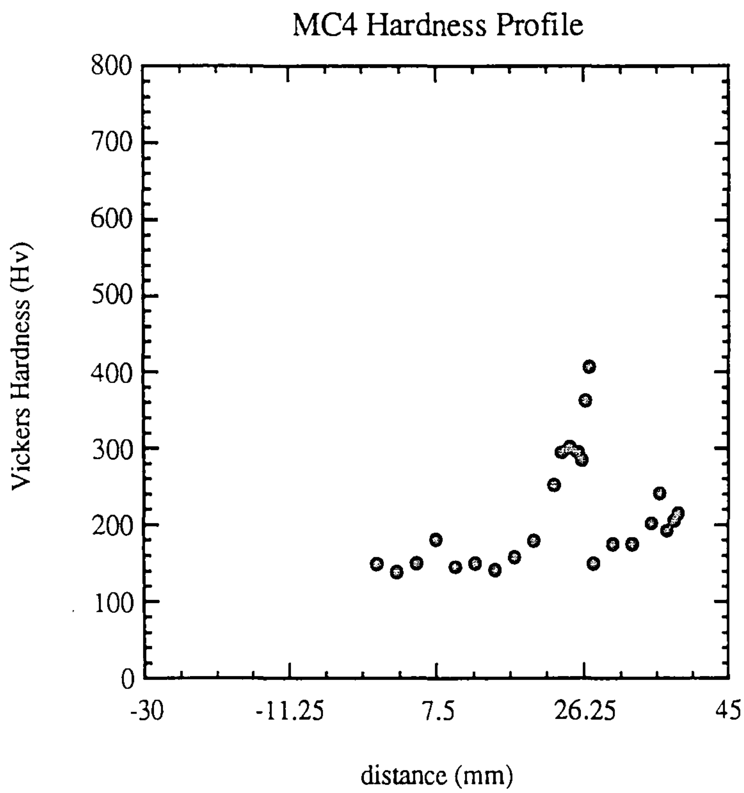


Figure 24 Hardness Profile

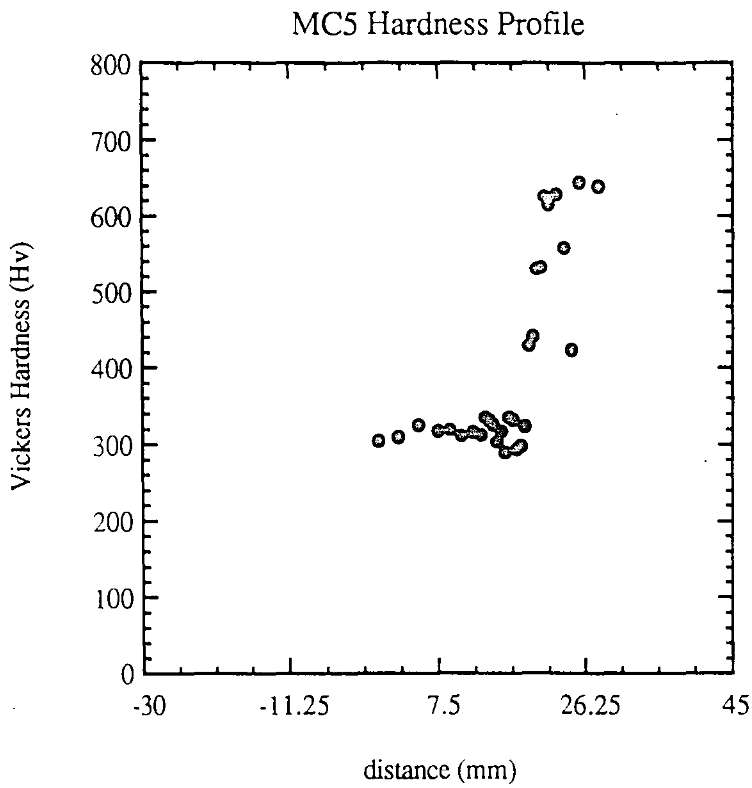


Figure 25 Hardness Profile

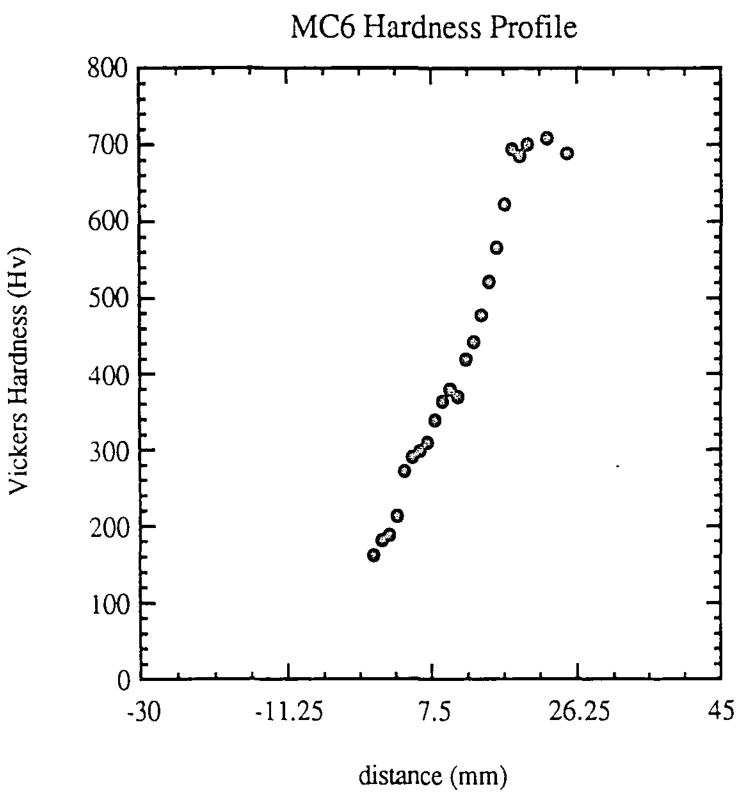


Figure 26 Hardness Profile

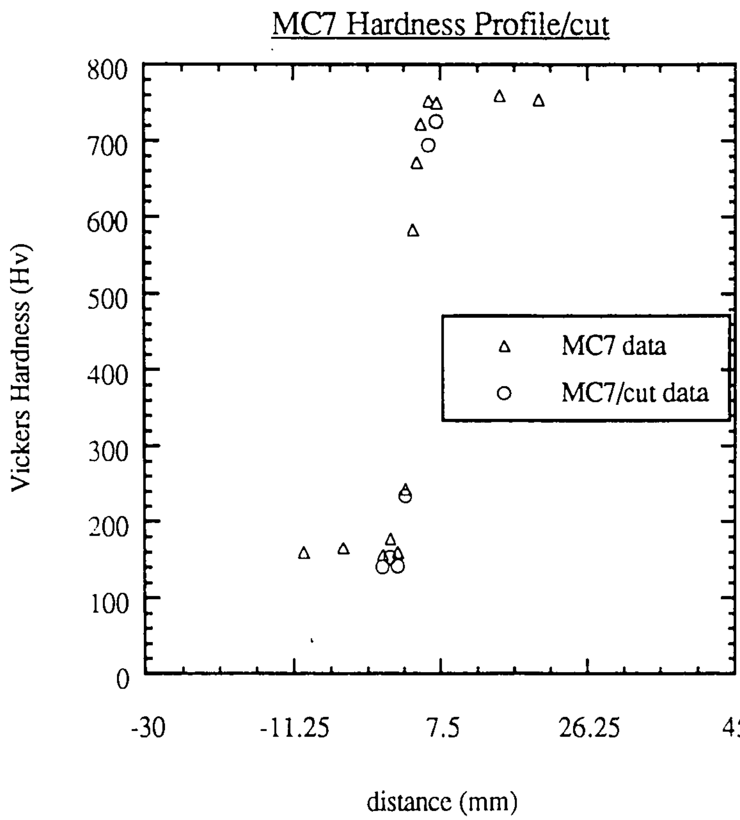


Figure 27 Hardness Profile

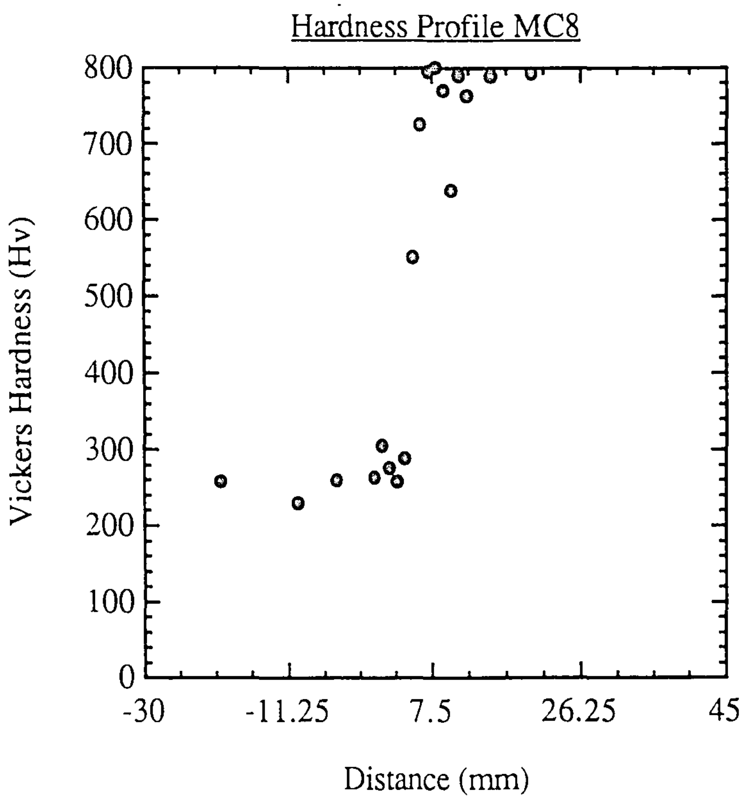


Figure 28 Hardness Profile

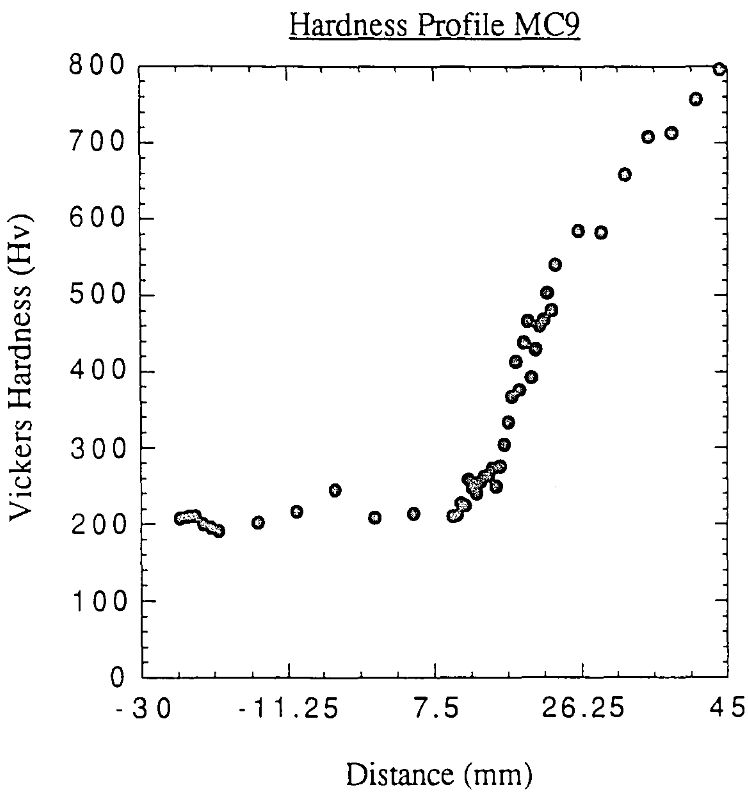


Figure 29 Hardness Profile

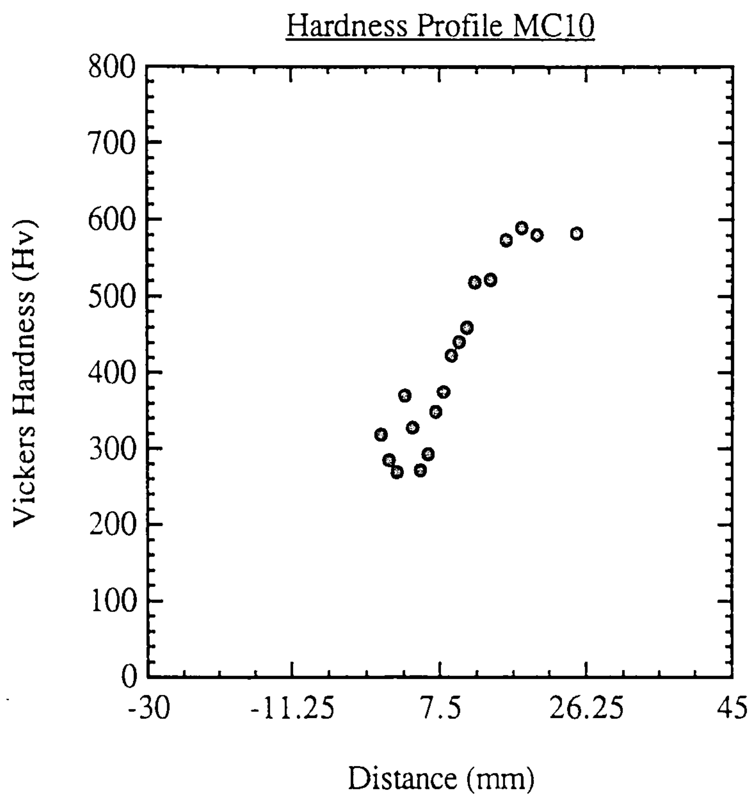


Figure 30 Hardness Profile

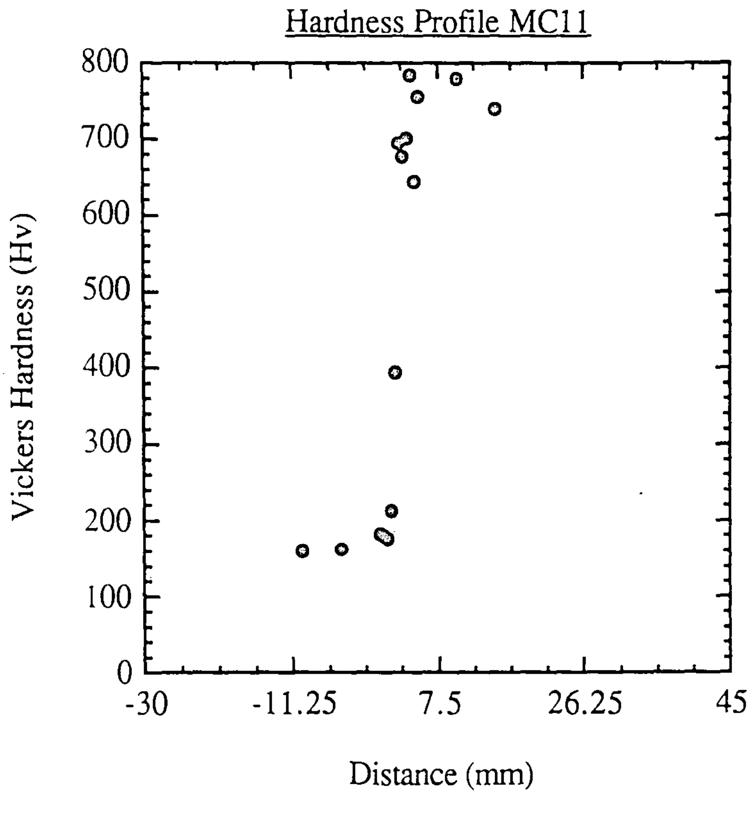


Figure 31 Hardness Profile

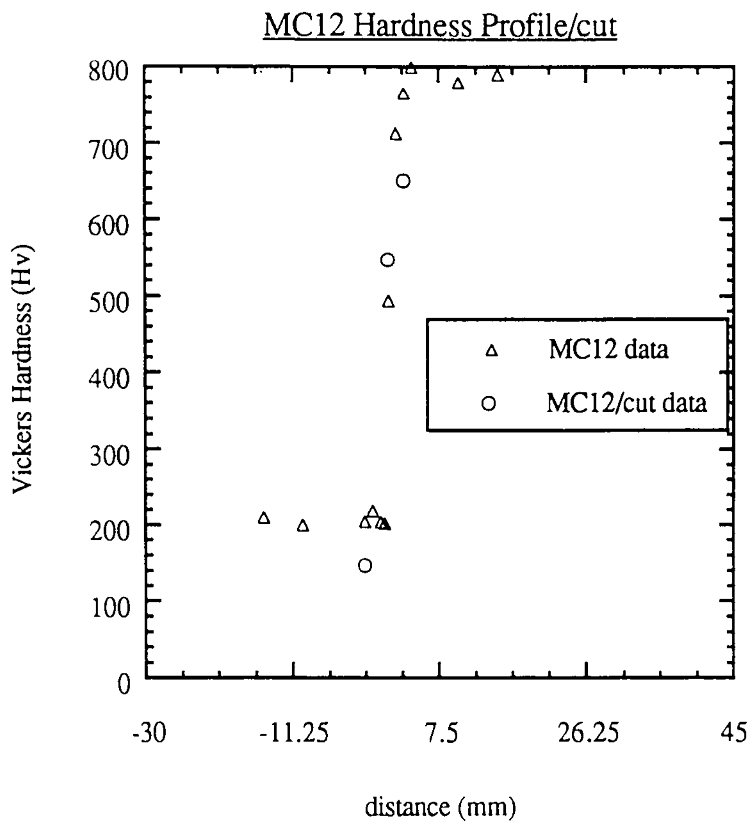


Figure 32 Hardness Profile

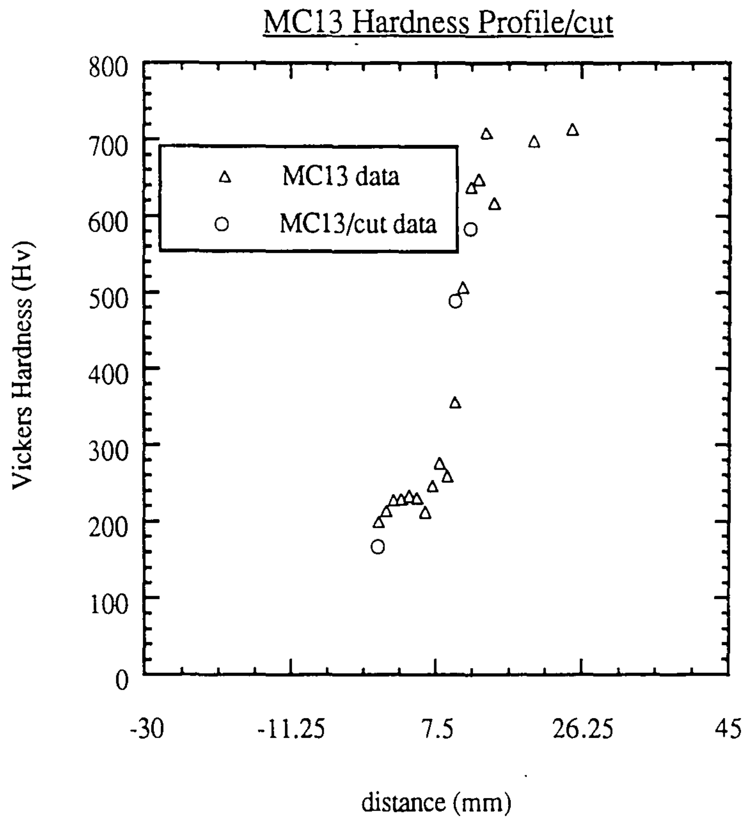


Figure 33 Hardness Profile

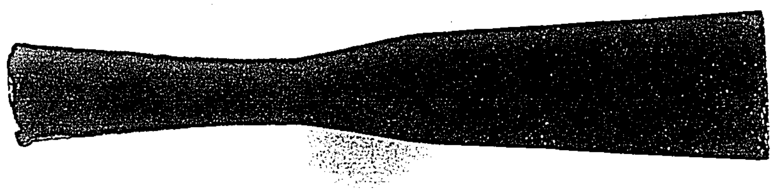
Table 5

Table for Socket and Tanged Chisels

sample	H-avg steel	H-avg iron	origin	maker	structure-st	structure-ir	date
SC1	625	325	B	R. Moore	M	F	1750
SC2	675	275	S	J. Green	M	F	1780
TC1	375	225	B	R. Moore	M	F	1750
TC3	175	175	S	P. Law	M	F	1770
TC4	725	375	S	?	M	F	1750

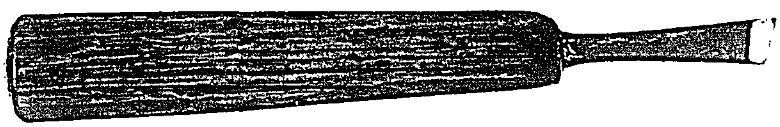
legend

M = martensite
F = ferrite
B = Birmingham
S = Sheffield
note : all hardness measurements have the units Hv



SC2

Figure 4



TC1

Figure 5

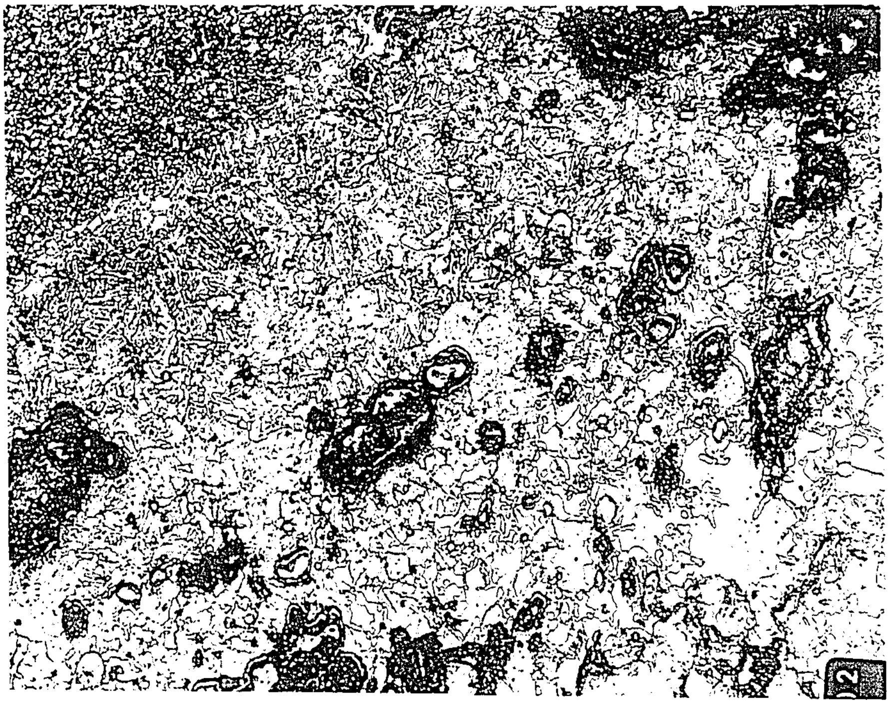


Figure 72 TC3, 160X, 4%picral

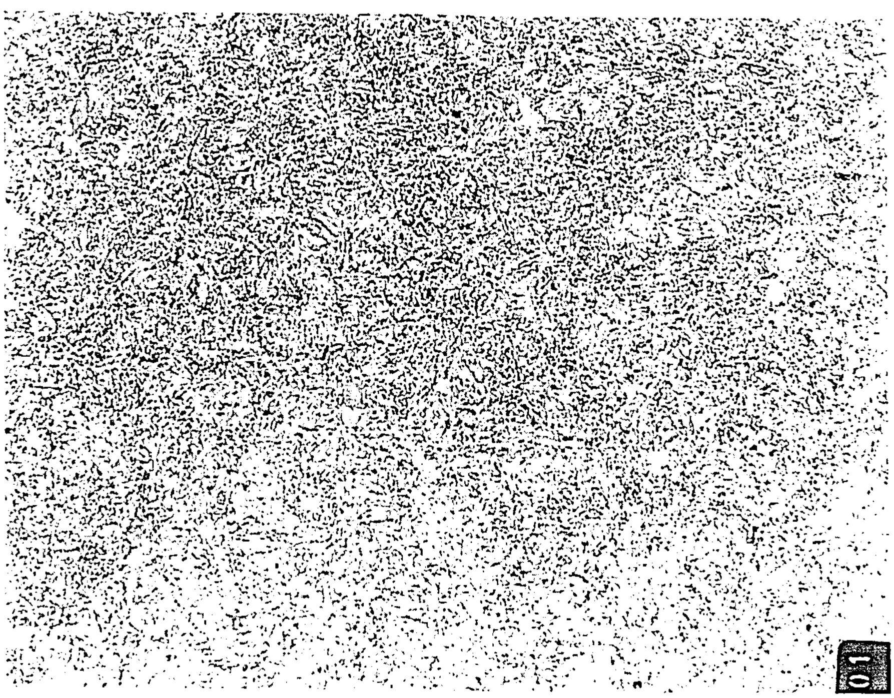


Figure 73 TC4, 500X, 4%picral

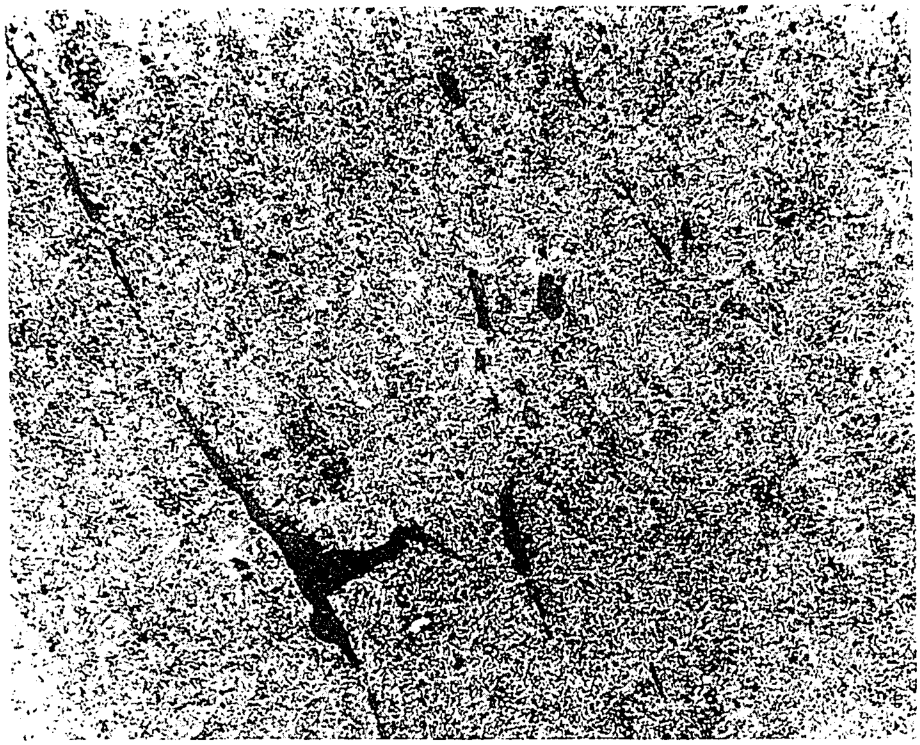


Figure 74 TC4, 600X, 4%picral

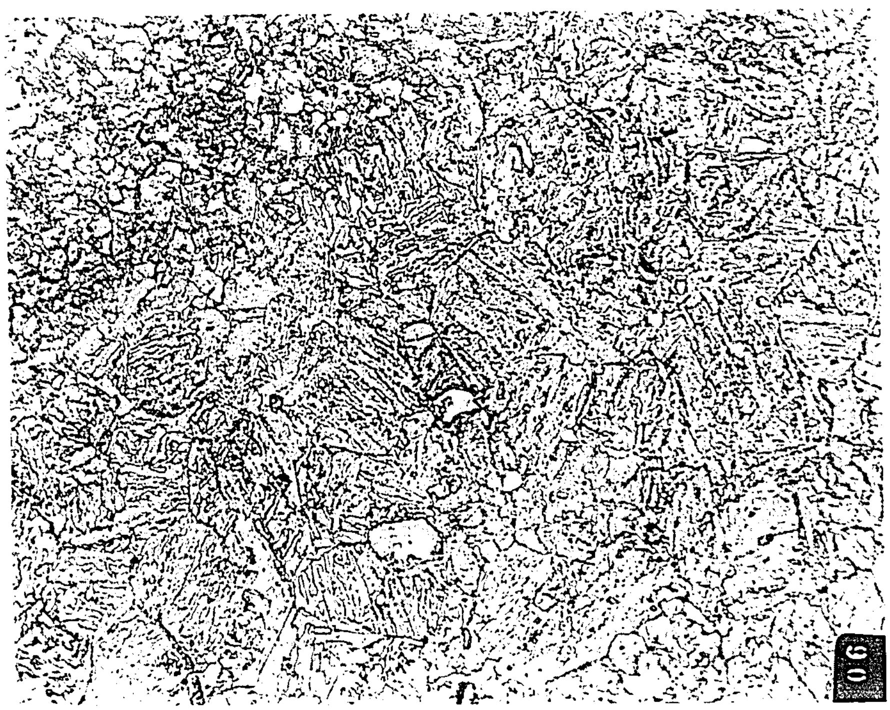


Figure 77 TC3, 400X, 4%picral

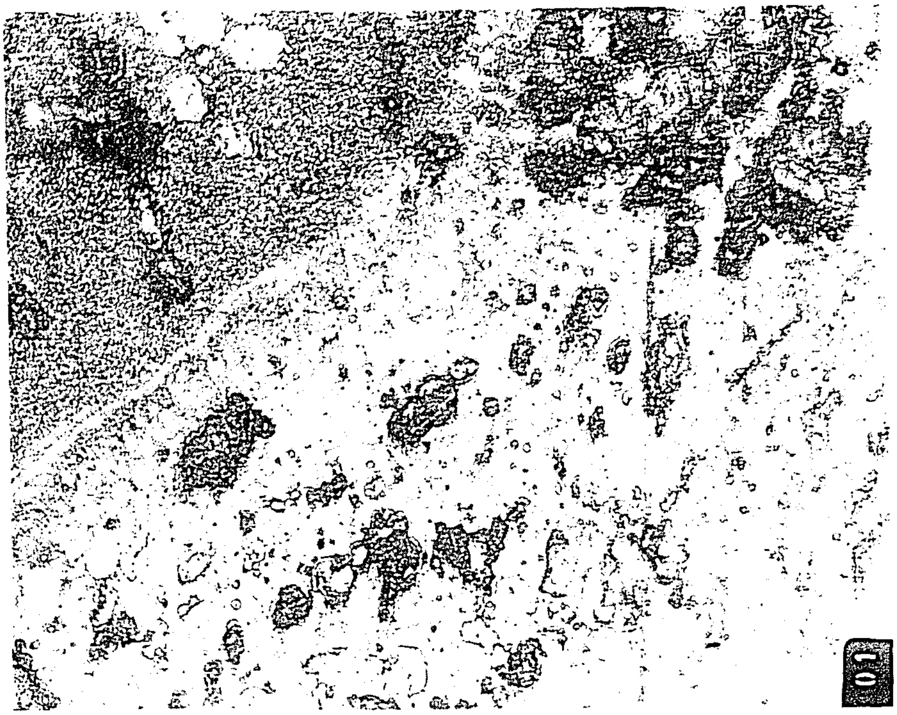


Figure 78 TC3, 100X, 4%picral

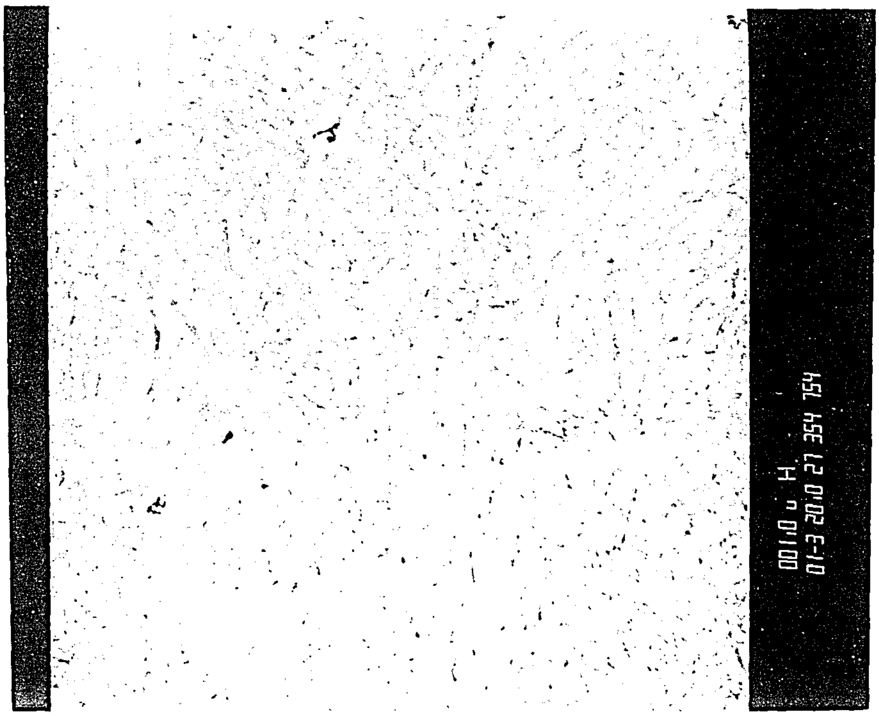


Figure 103 TC4, 1000X, 4%picral, SE image

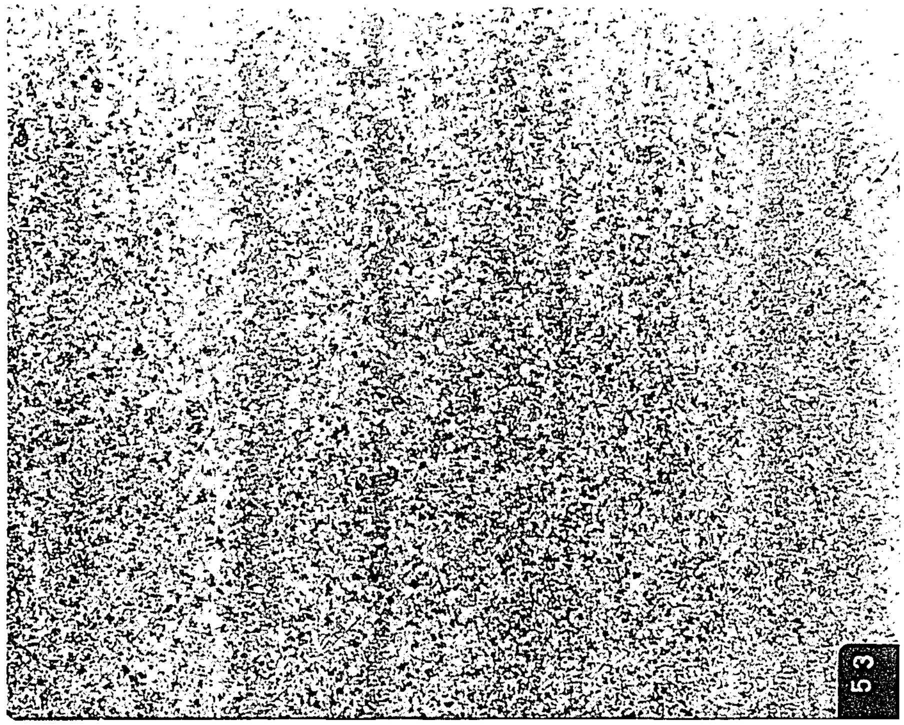


Figure 80 SC1-tip, 60X, 4%picral

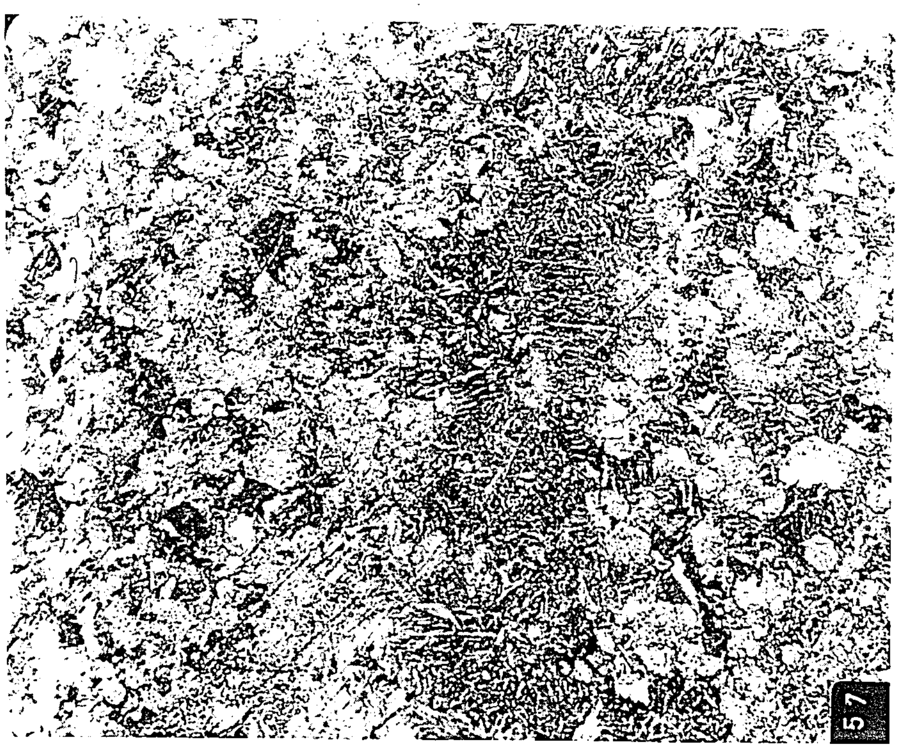


Figure 81 SC1-tip, 500X, 4%picral

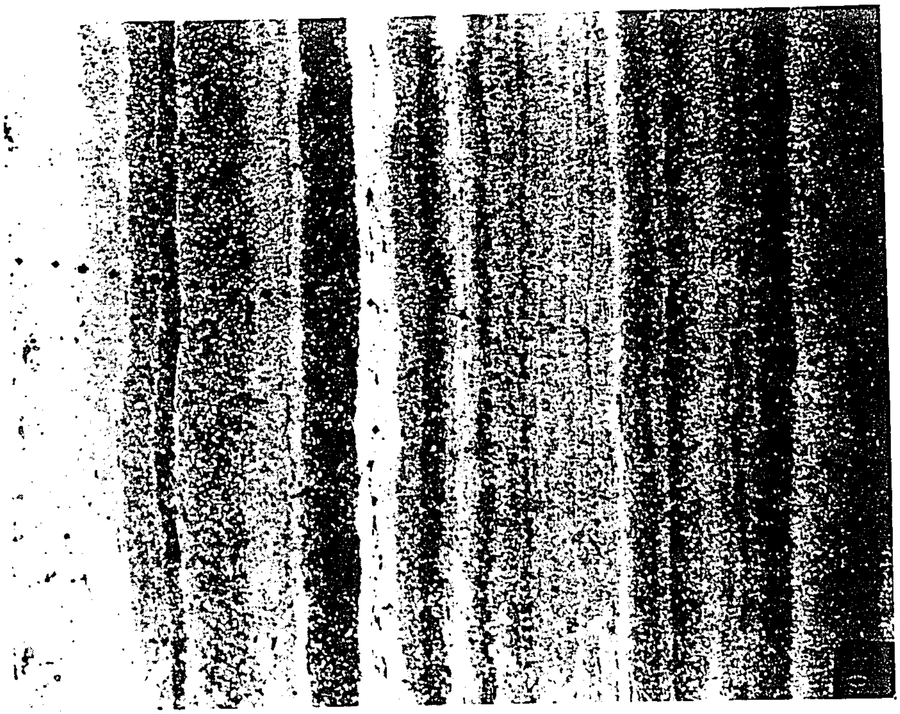


Figure 82 SC1-tip, 50X, Oberhoffer's

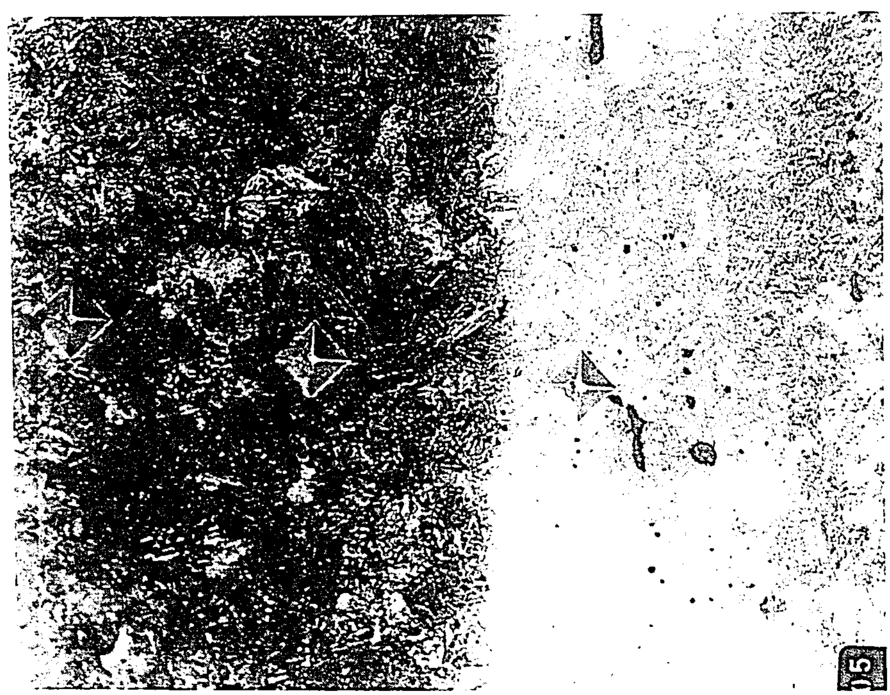


Figure 83 SC1-tip, 800X, Oberhoffer's

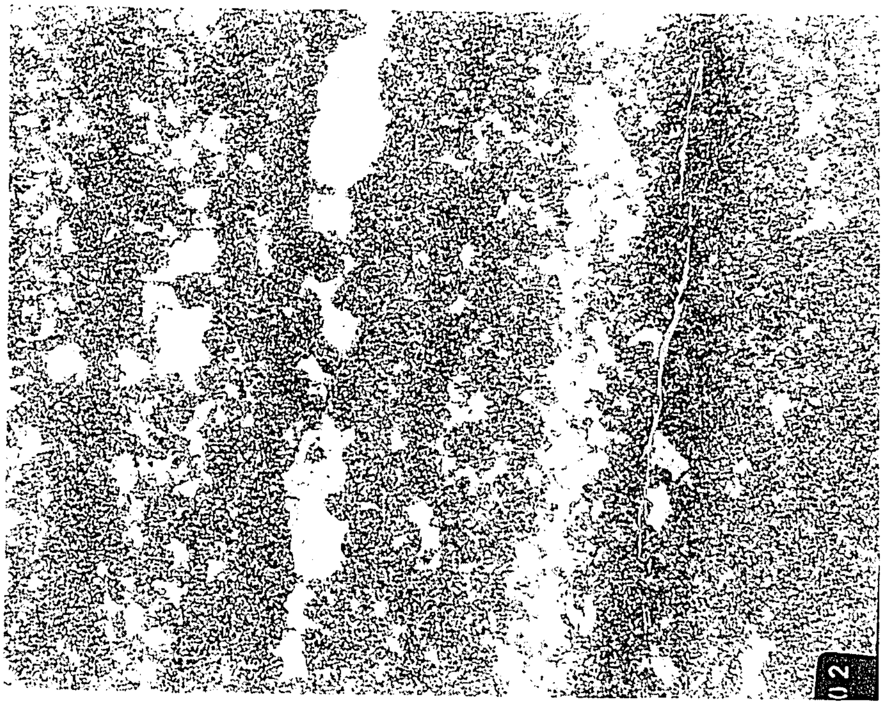


Figure 84 SC1-tip, 125X, Stead's

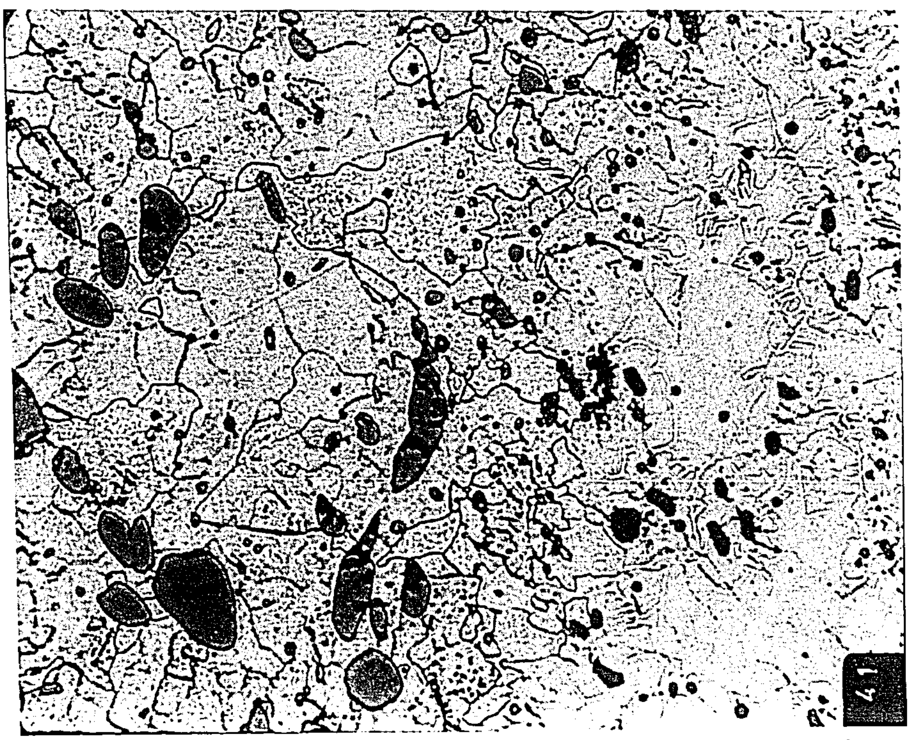


Figure 85 SC1-tip, 200X, 4%picral

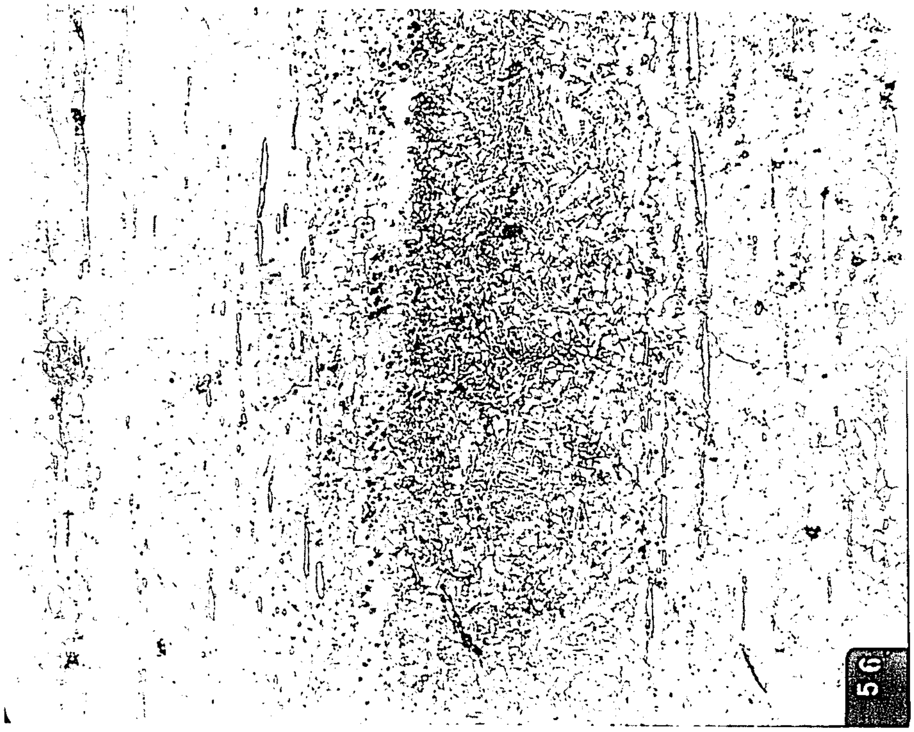


Figure 86 SC1-socket, 65X, 4%picral

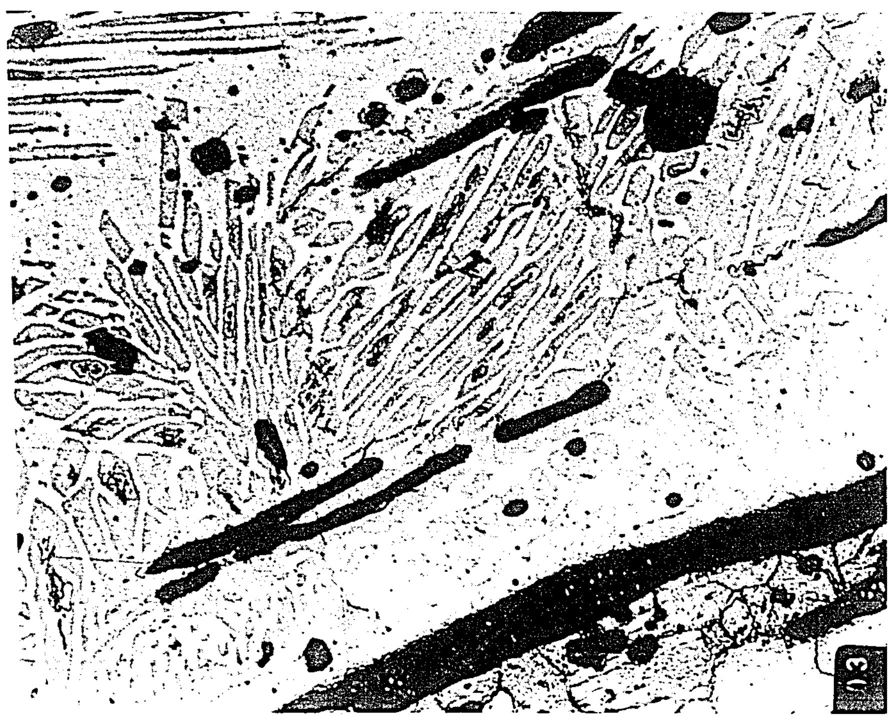


Figure 87 SC1-socket, 320X, Stead's

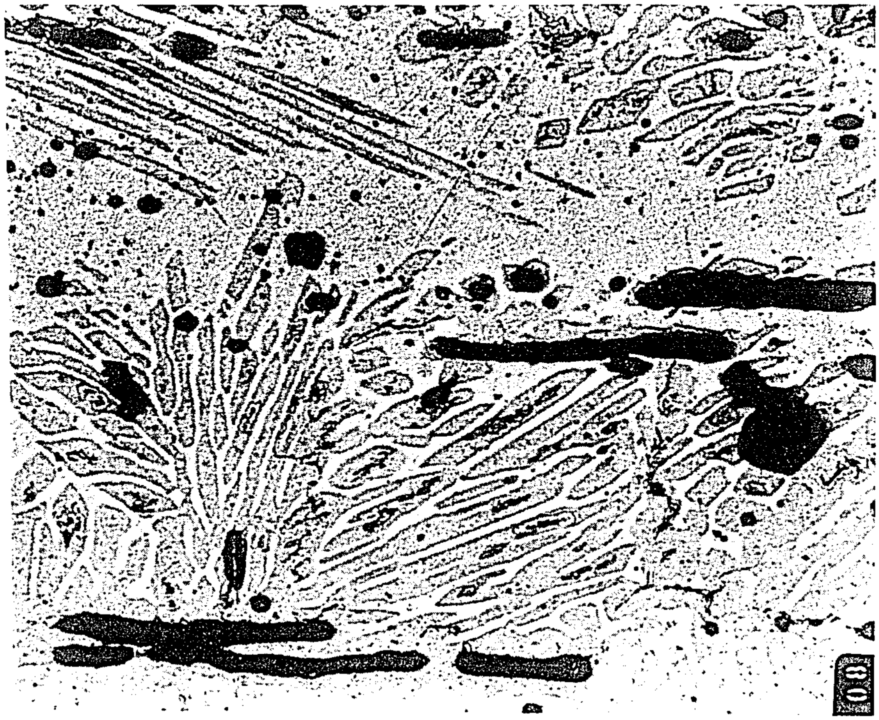


Figure 88 SC1-socket, 320X, Oberhoffer's

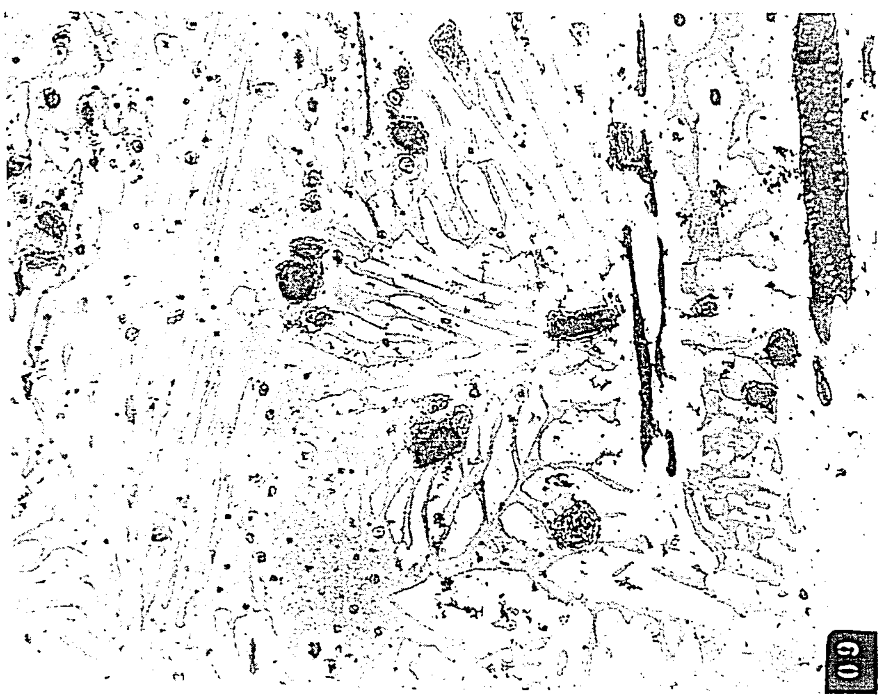


Figure 89 SC1-socket, 310X, 4%picral

Figure 90 SC1-soc, 4%picral, SE image

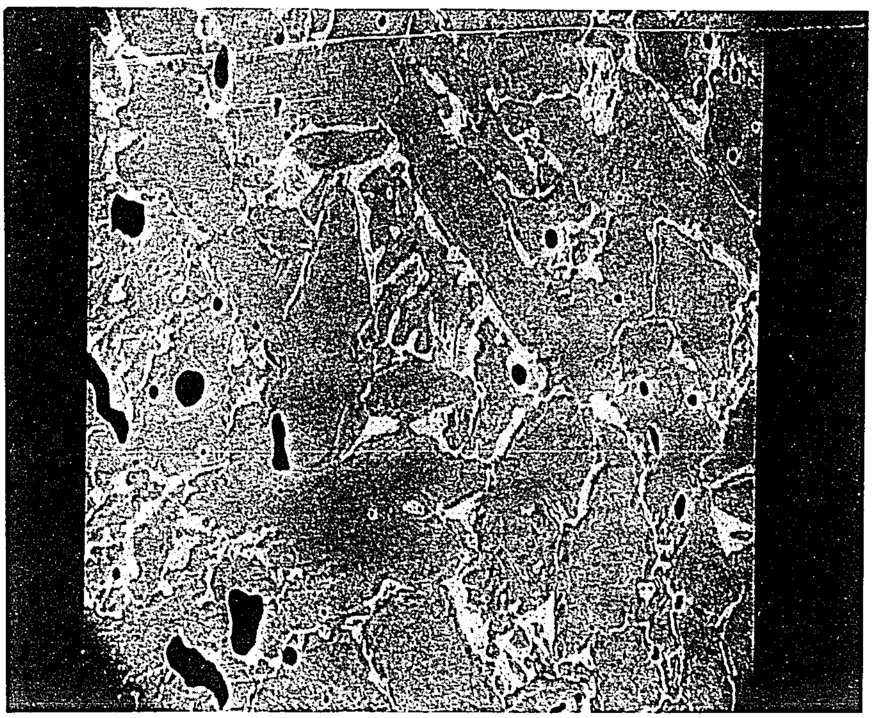


Figure 91 SC1-soc, 1000X, 4%picral

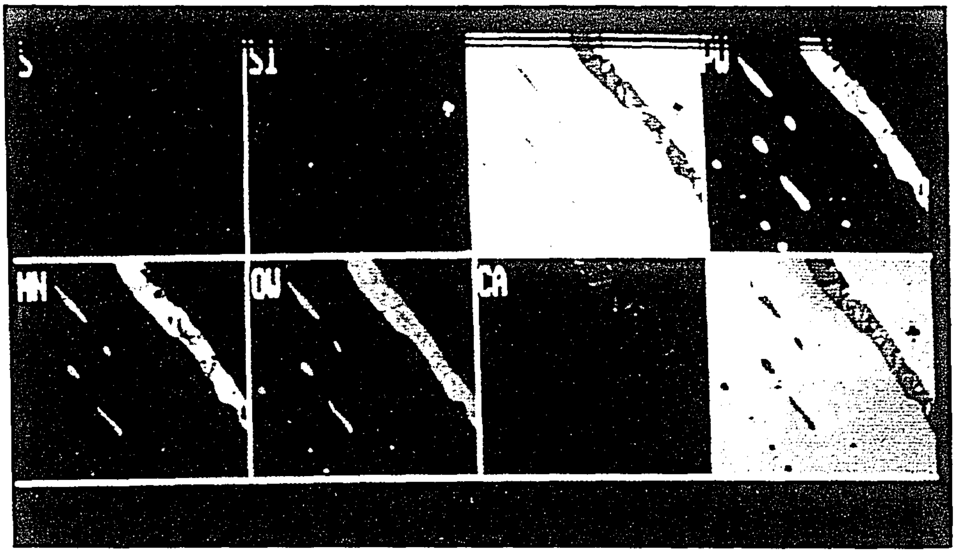


Figure 93 Xray map from SC1-socket

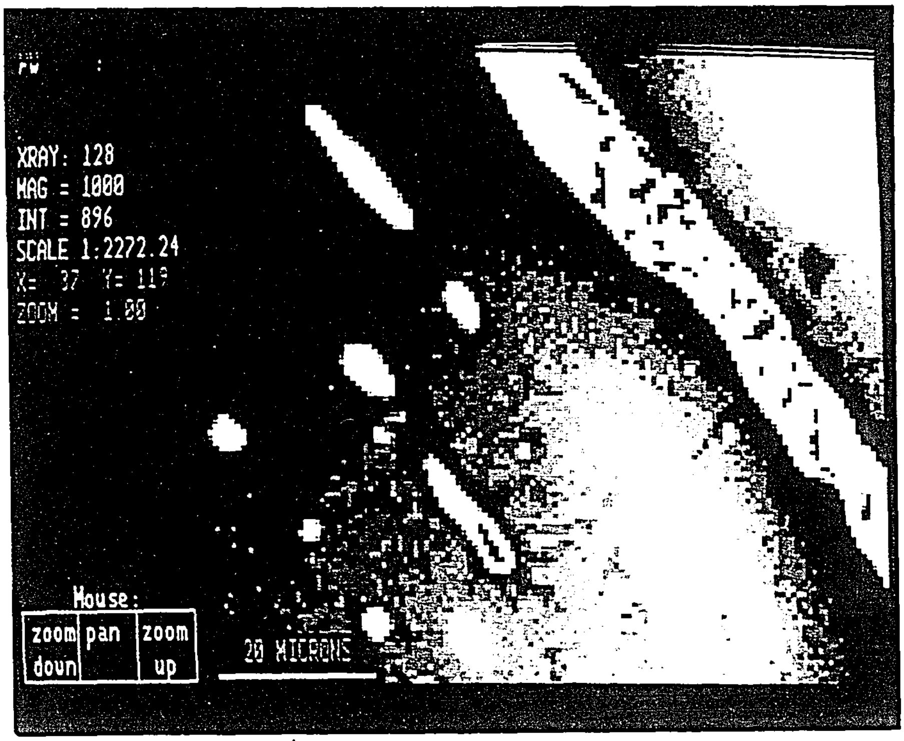


Figure 92 Xray for P from SC1-socket

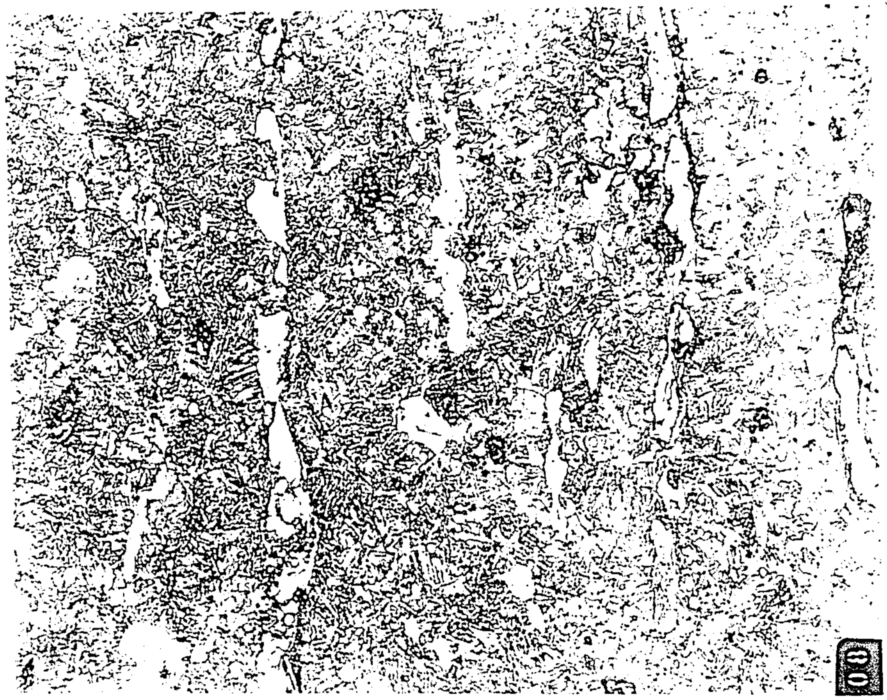


Figure 96 SC2, 160X, 4%picral

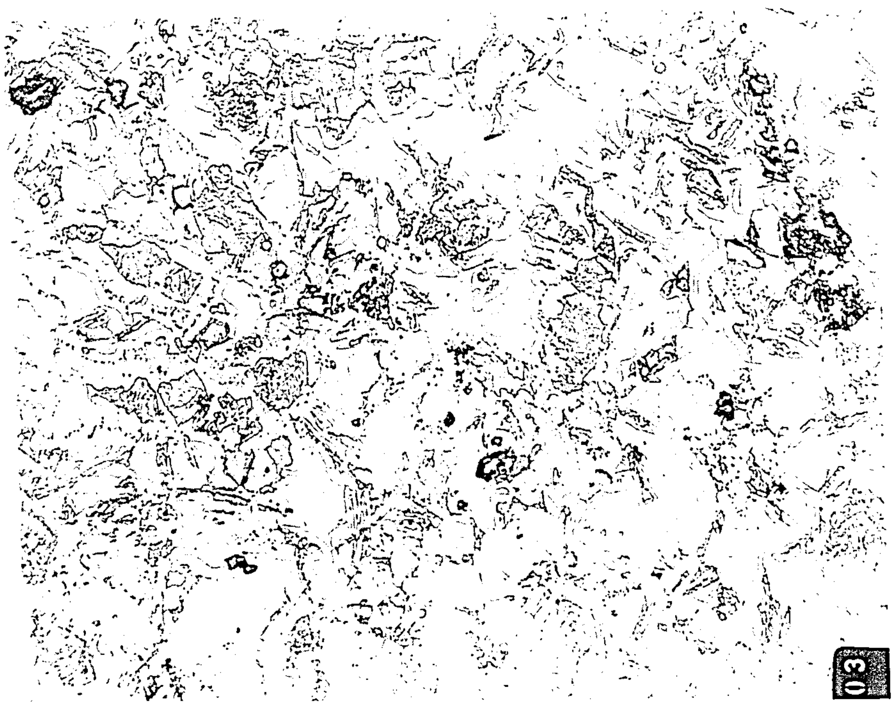


Figure 97 SC2, 400X, 4%picral

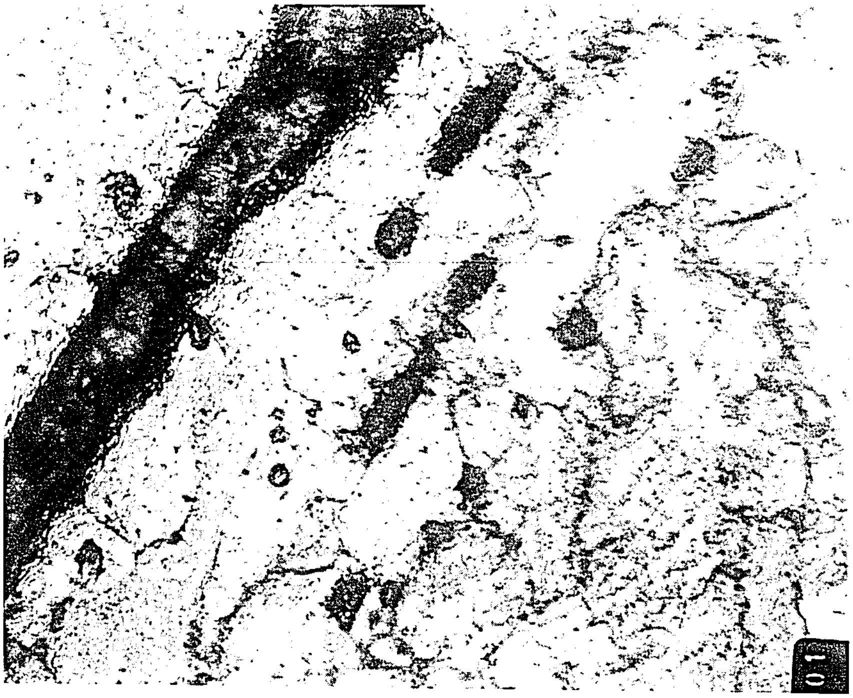


Figure 98 SC1-socket, 800X, Stead's

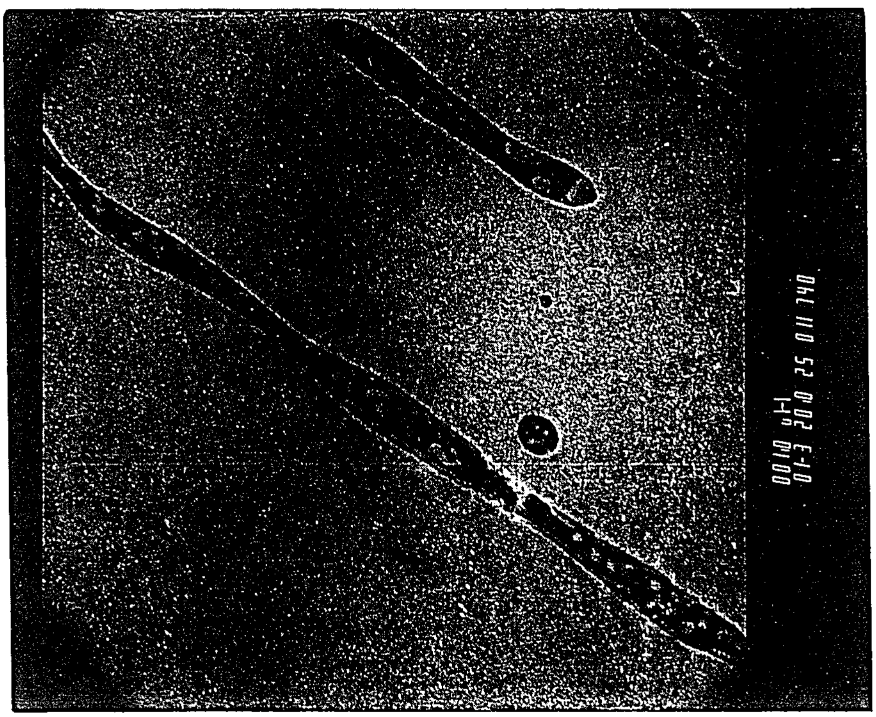


Figure 99 SC1-socket, 4%picral, SE image

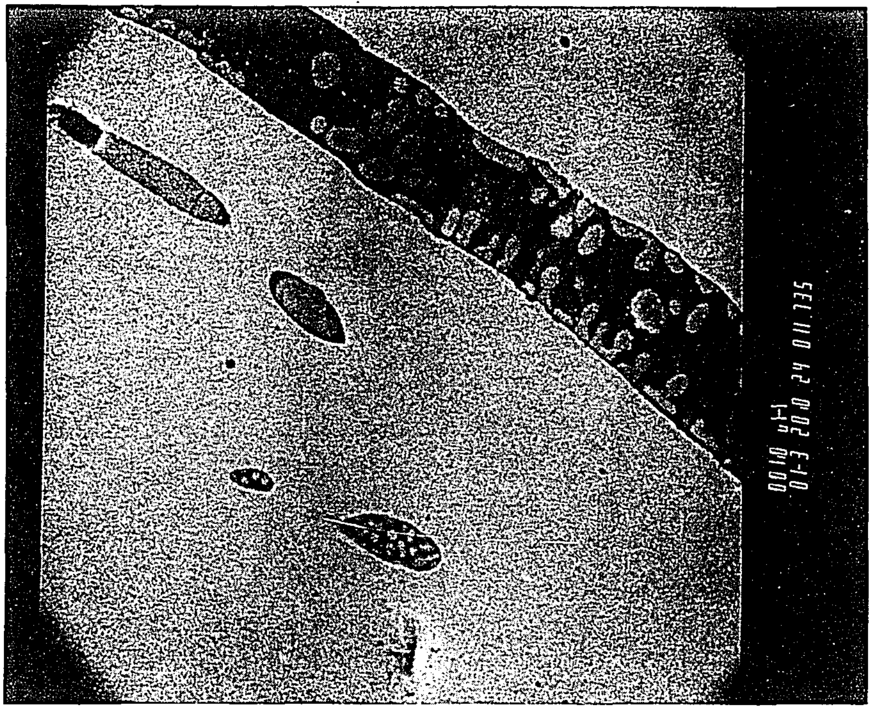


Figure 100 SC1-socket, 1000X, 4%picral, SE image

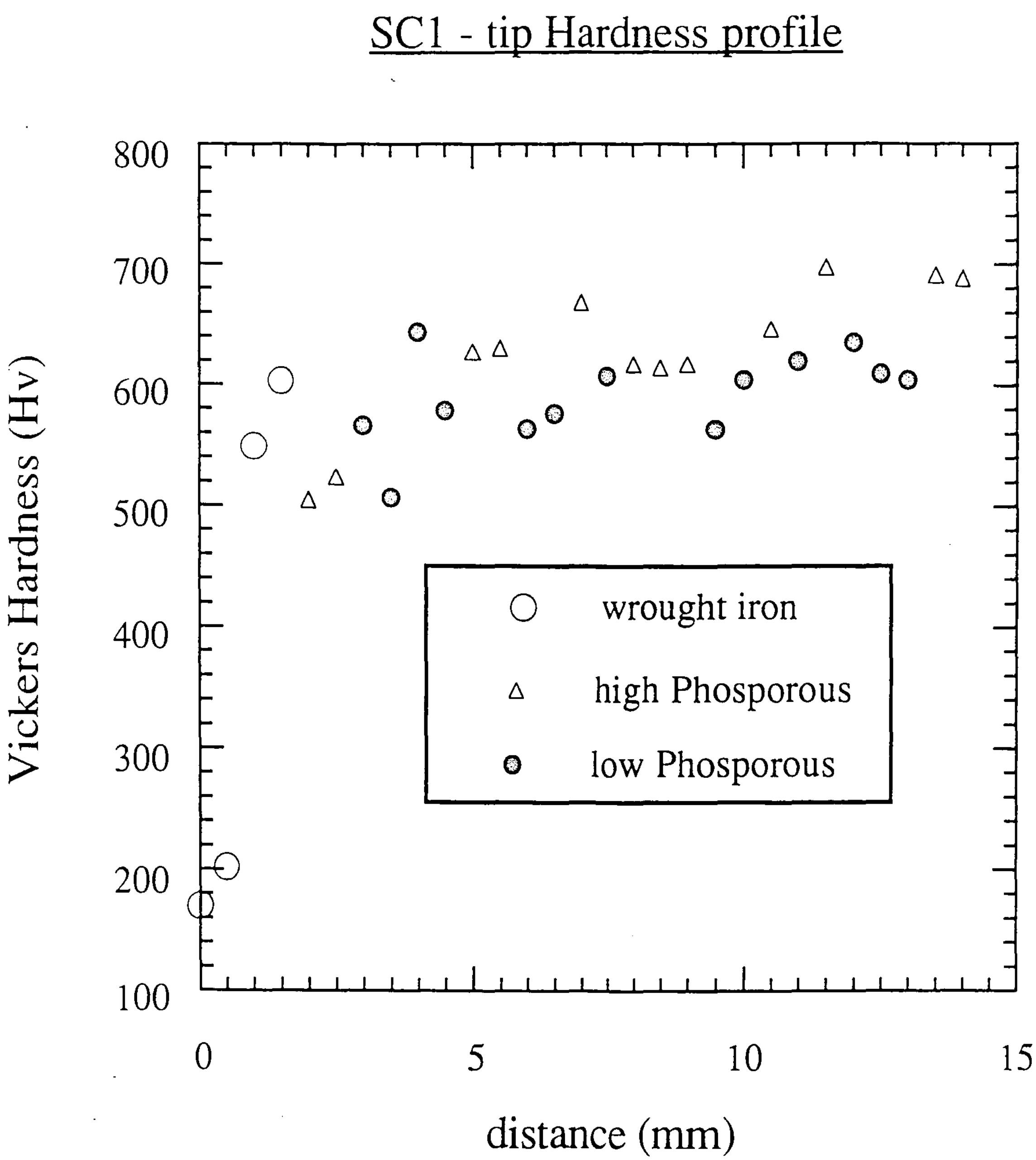


Figure 94 Hardness Profile

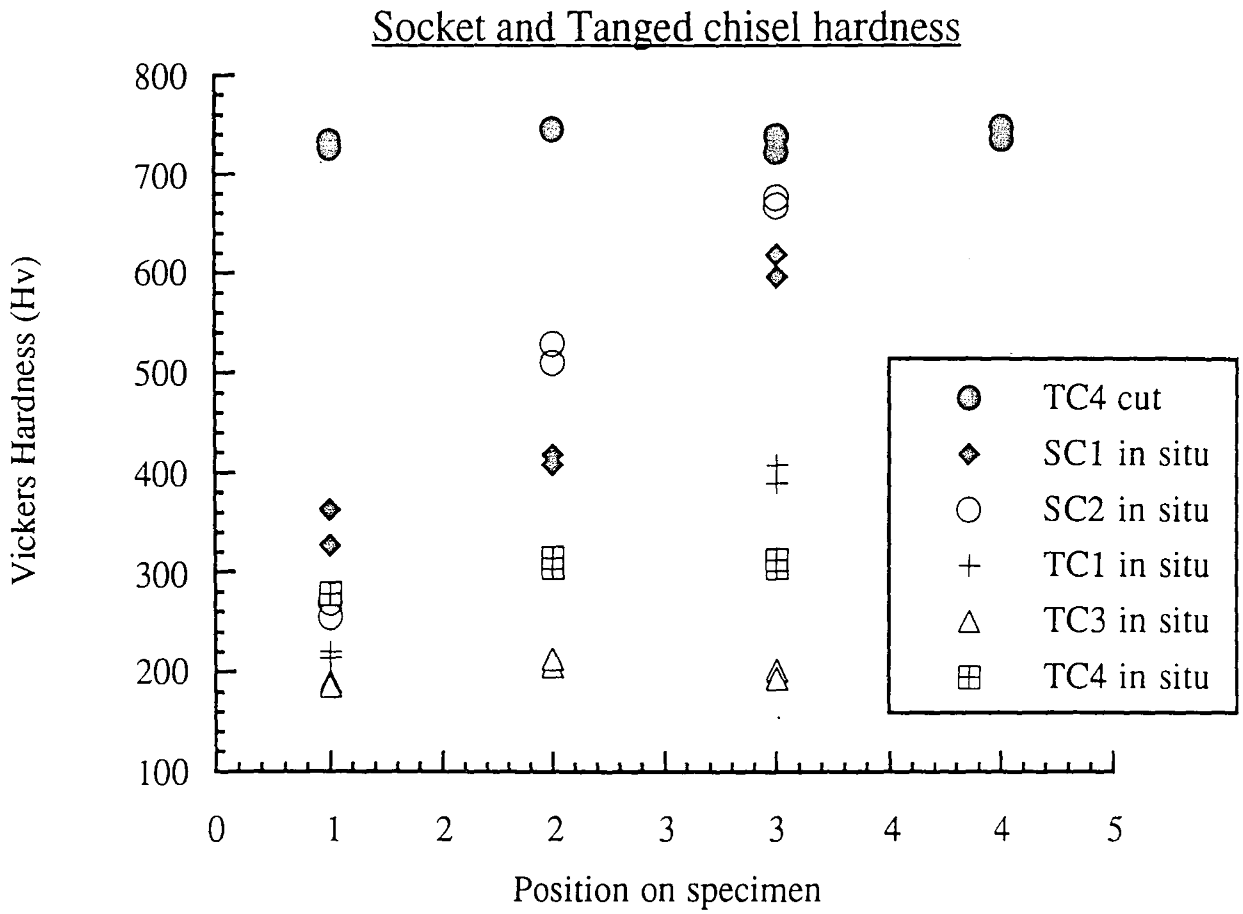


Figure 79 Hardness for socket and tanged chisels

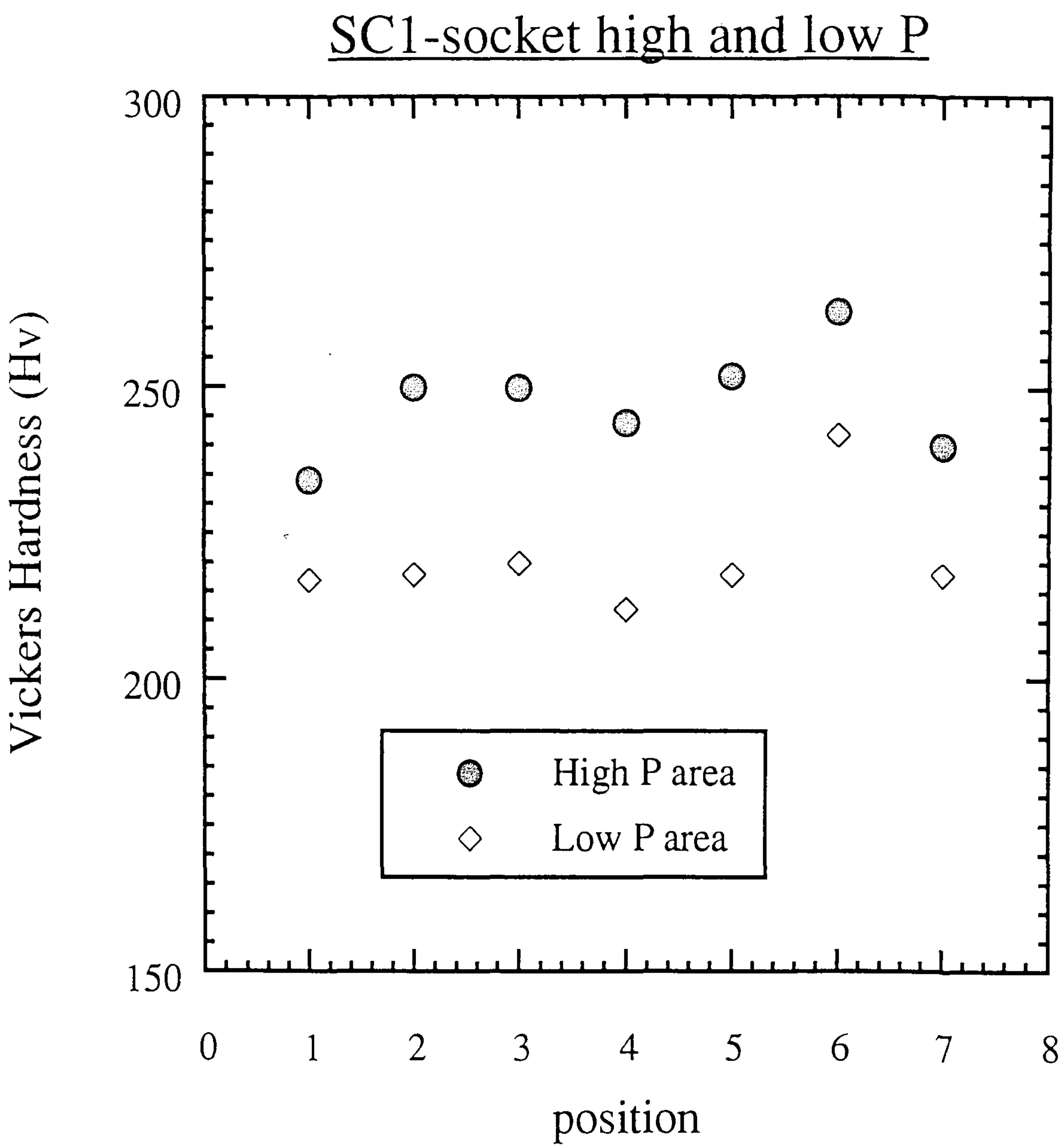


Figure 95 Hardness from SC1-socket

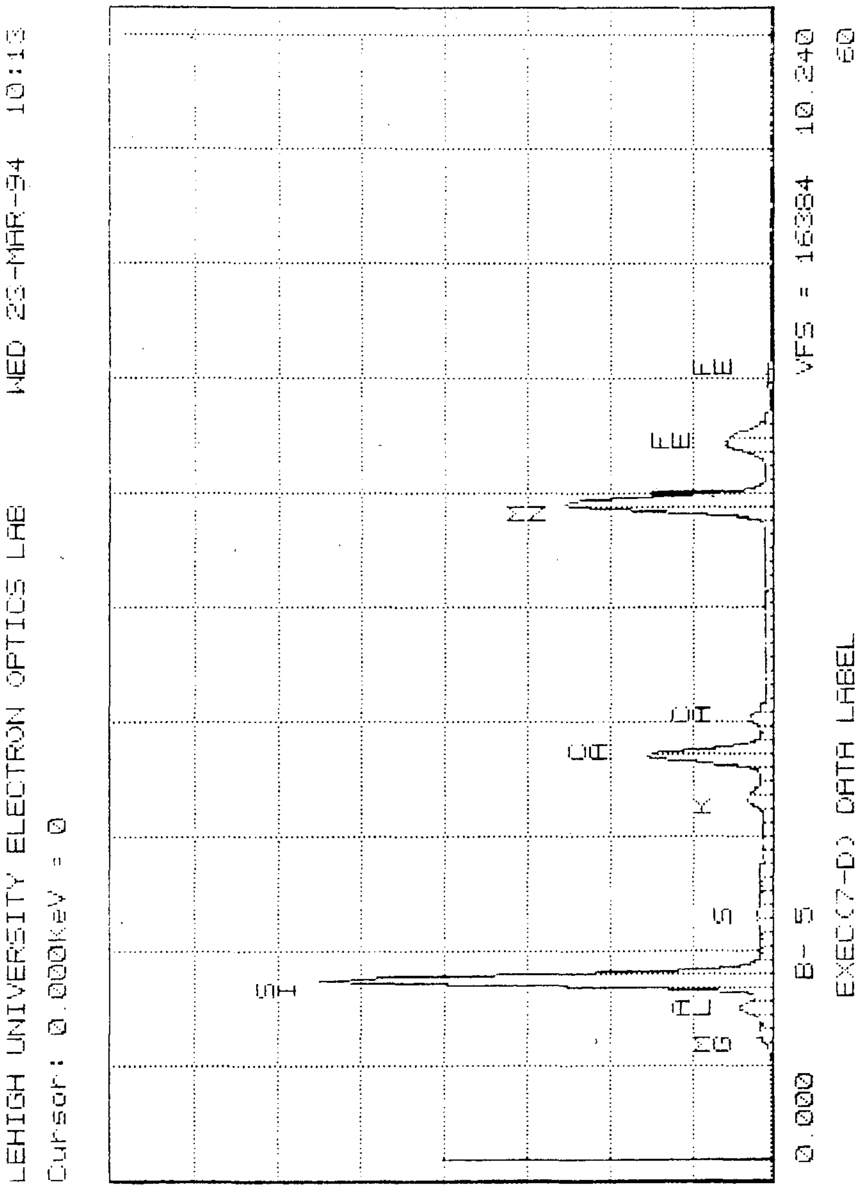


Figure 75 Spectra from TC4 "fold" slag (EDS)

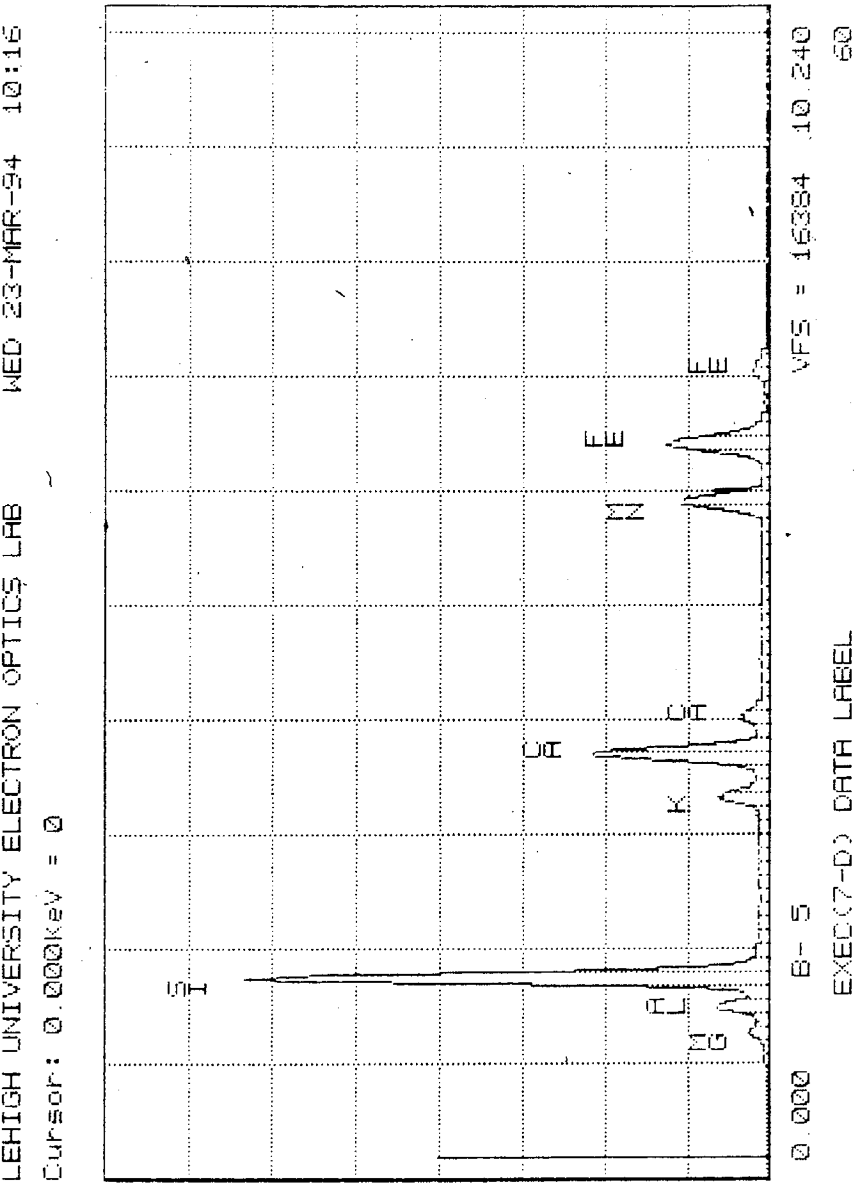


Figure 76 Spectra from TC4 slag (EDS)

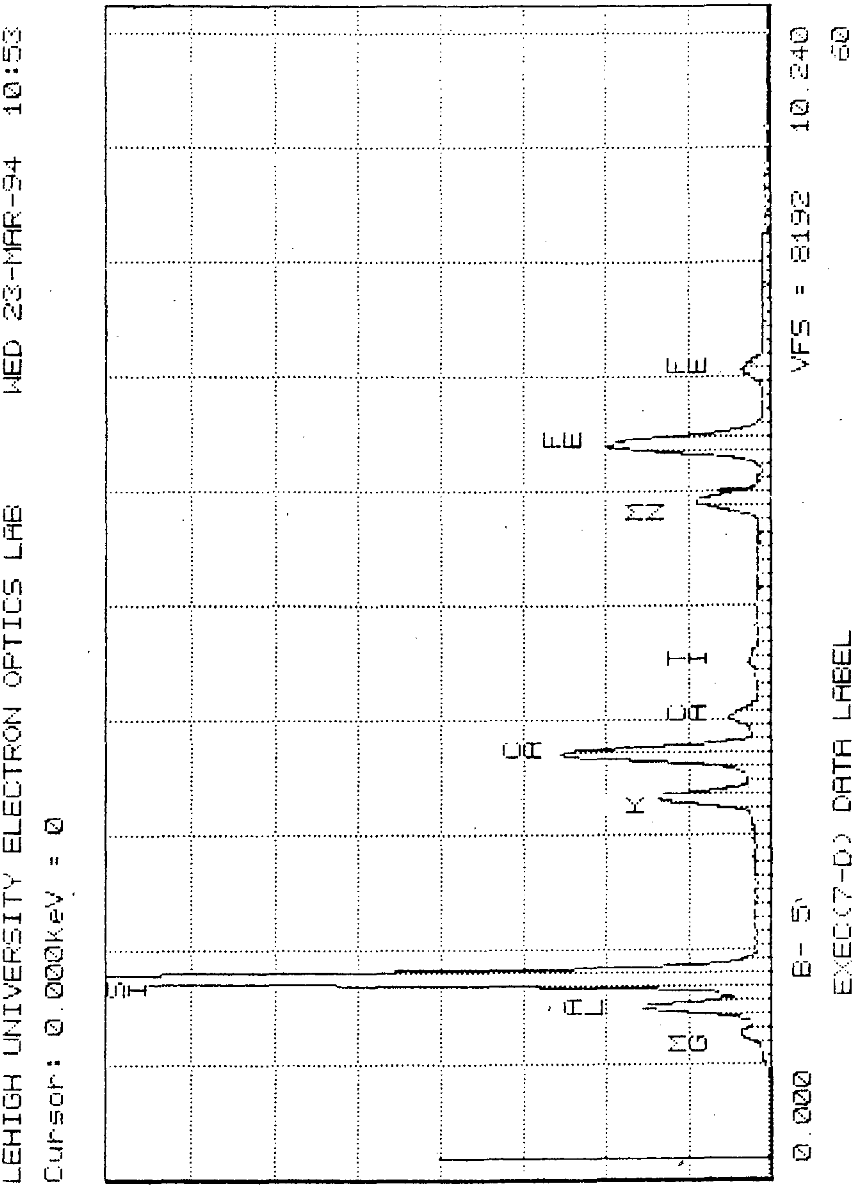


Figure 34 Spectra from SC1-tip, slag (EDS)

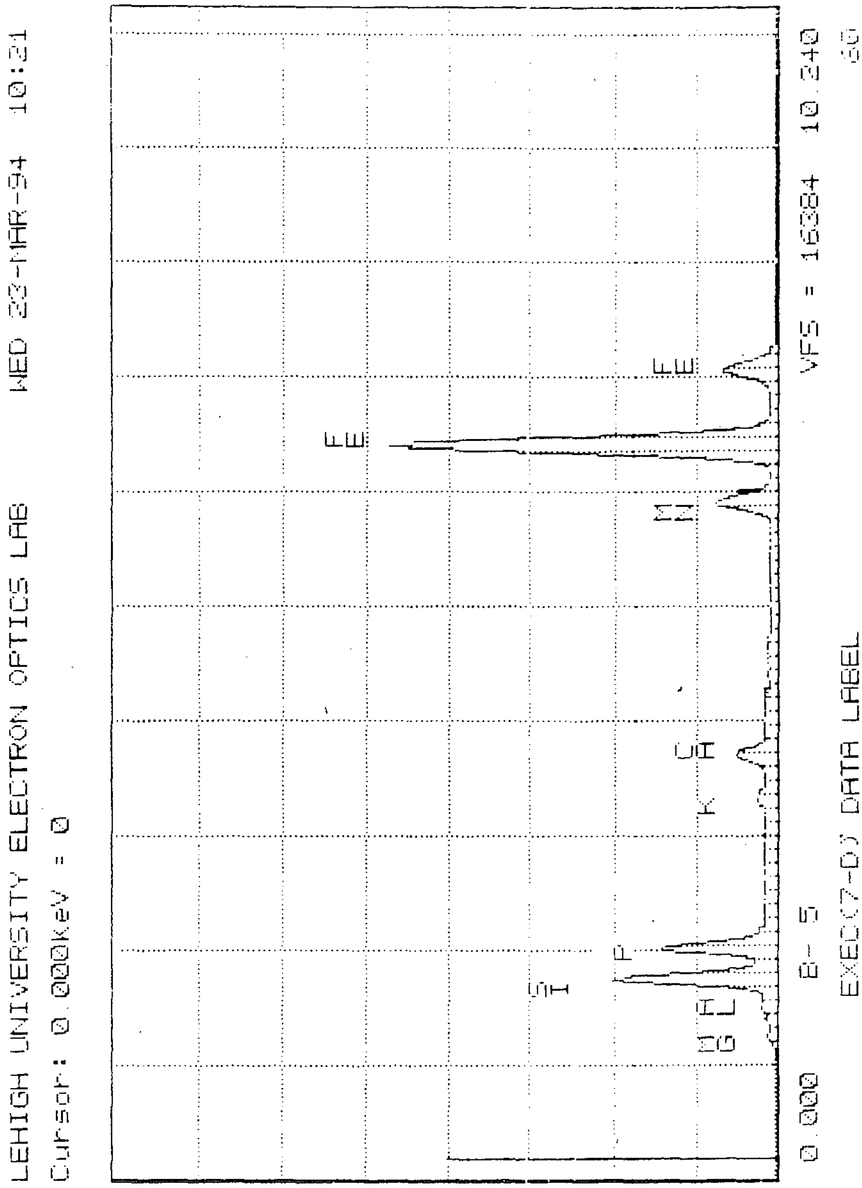


Figure 35 Spectra from SC1-socket, 2 phase slag (EDS)

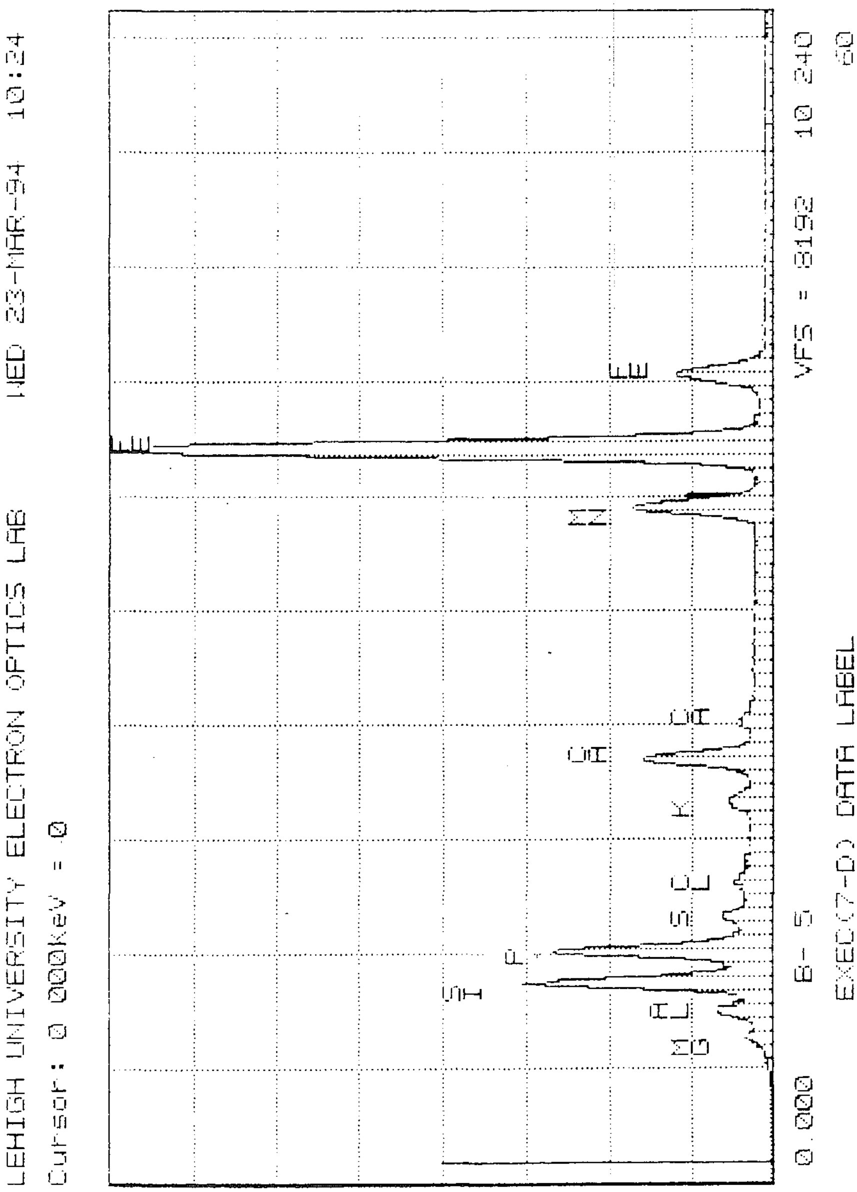


Figure 36 Spectra from SC1-socket, one phase slag (EDS)

Table 2

Table for Saws

sample	H-avg steel	H-avg backing	origin	structure-st	structure-ir	date	maker
S1	425	125	B	M	70/30 BR	1767	Dalaway
S2	575	125	L	M	70/30 BR	1760	Manwaring
S3	450	125	LS	M	70/30 BR	1787	?
S4	450	125	S	M	F	1787	Kenyon
S5	500	125	L	M	70/30 BR	?	?
S6	360	125	S	M	70/30 BR	1816	Spear

legend

M = martensite
F = ferrite
BR = brass
B = Birmingham
L = London
LS = London Spring
S = Sheffield
note : all hardness measurements have the measurements Hv

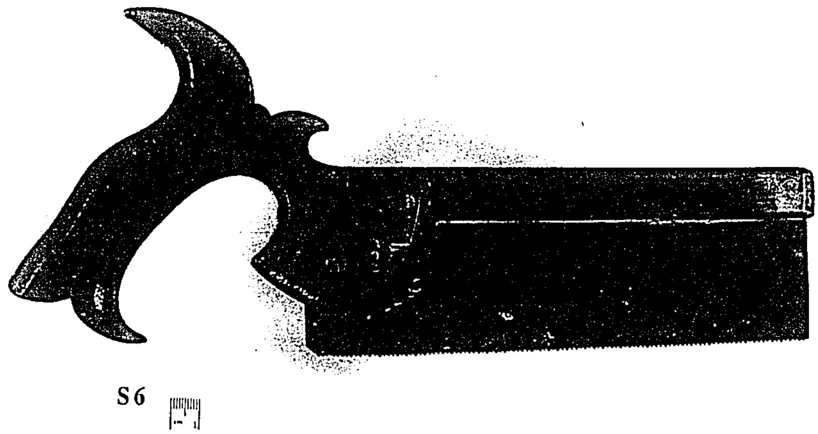


Figure 1

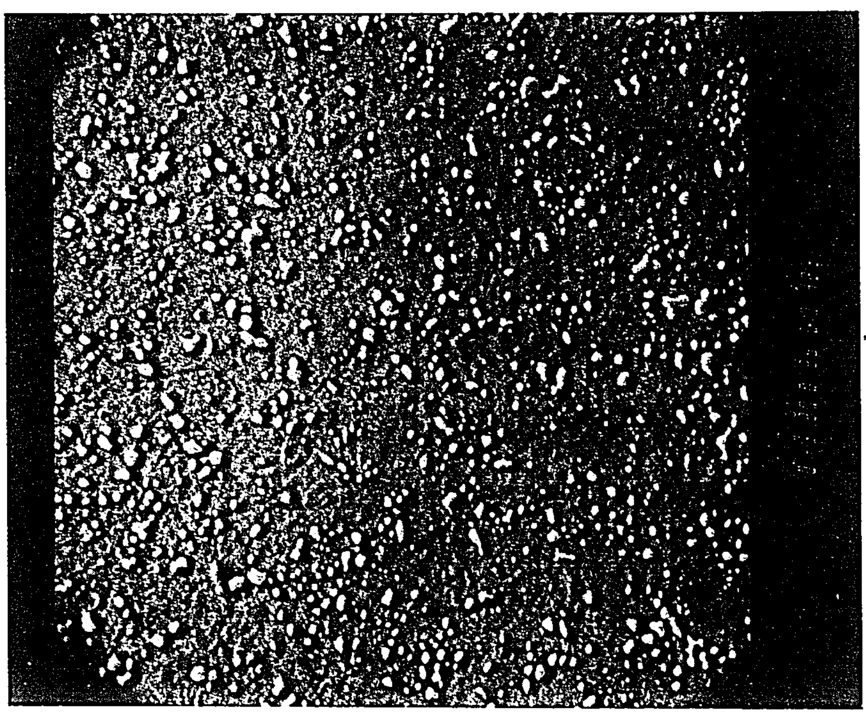


Figure 61 S2, 1000X, 4%picral, SE image

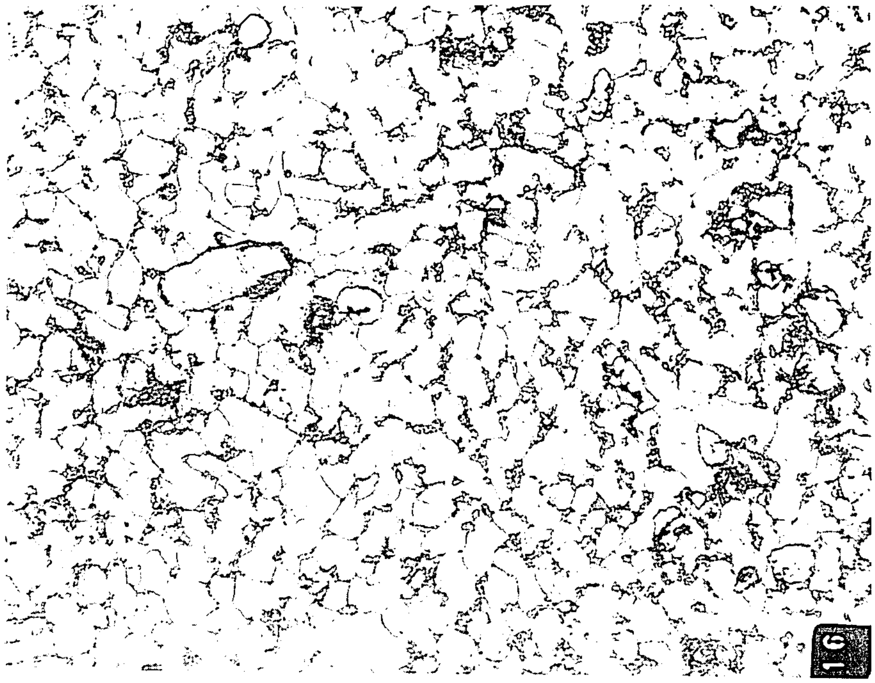


Figure 62 S4 backing, 300X, 2%nital

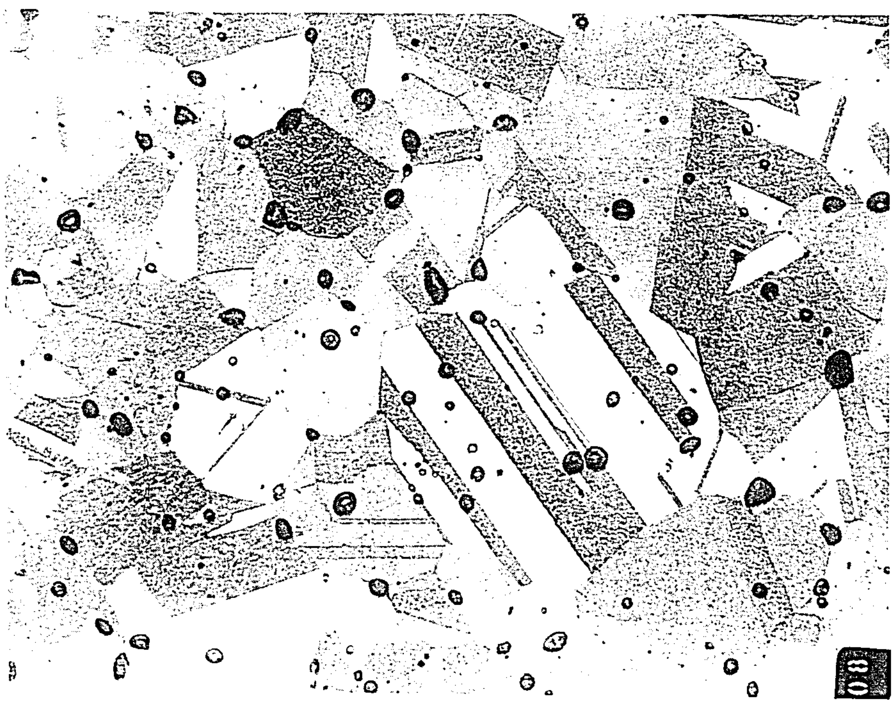


Figure 66 S3 backing, 200X, brass etch

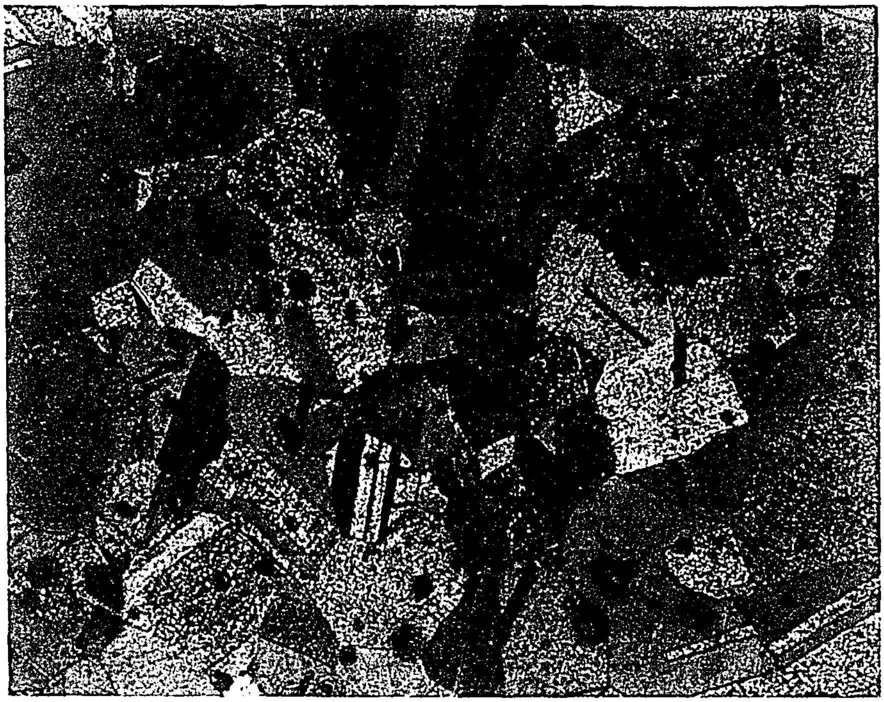


Figure 67 S3 backing, 150X, brass etch, polarized light

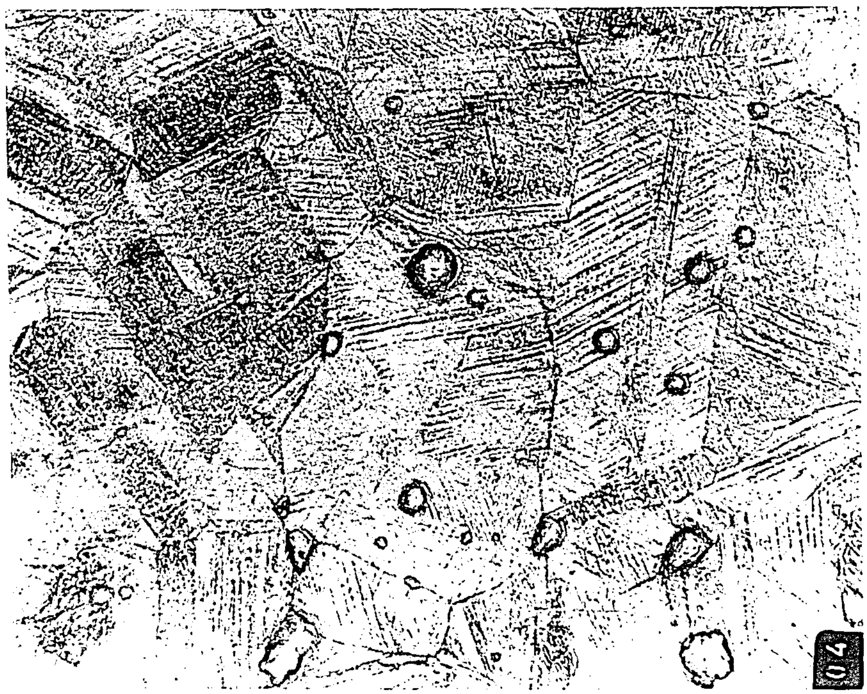


Figure 68 S1 backing, 500X, brass etch

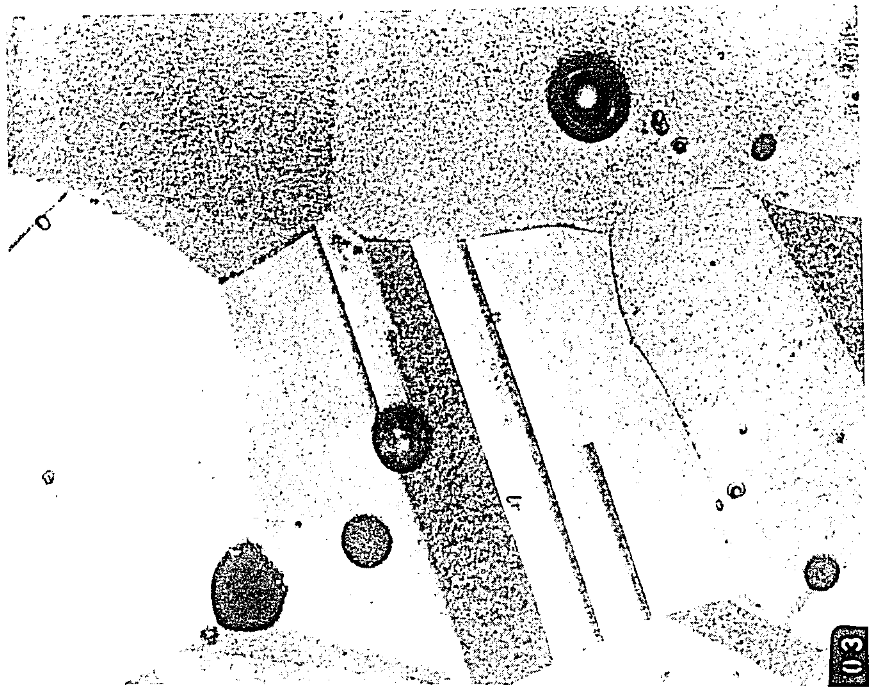


Figure 69 S3 backing, 800X, brass etch

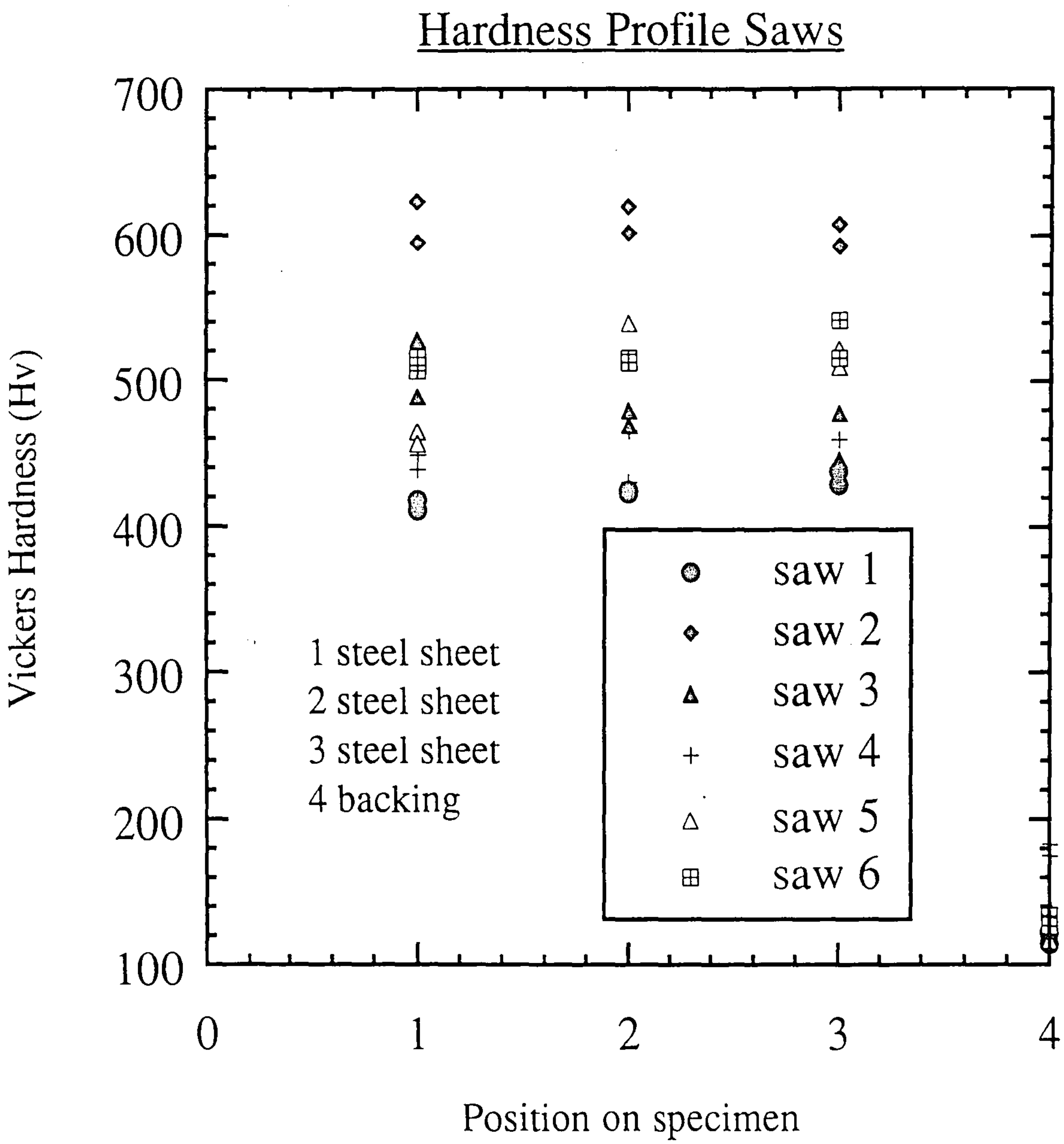


Figure 64 Hardness of Saws

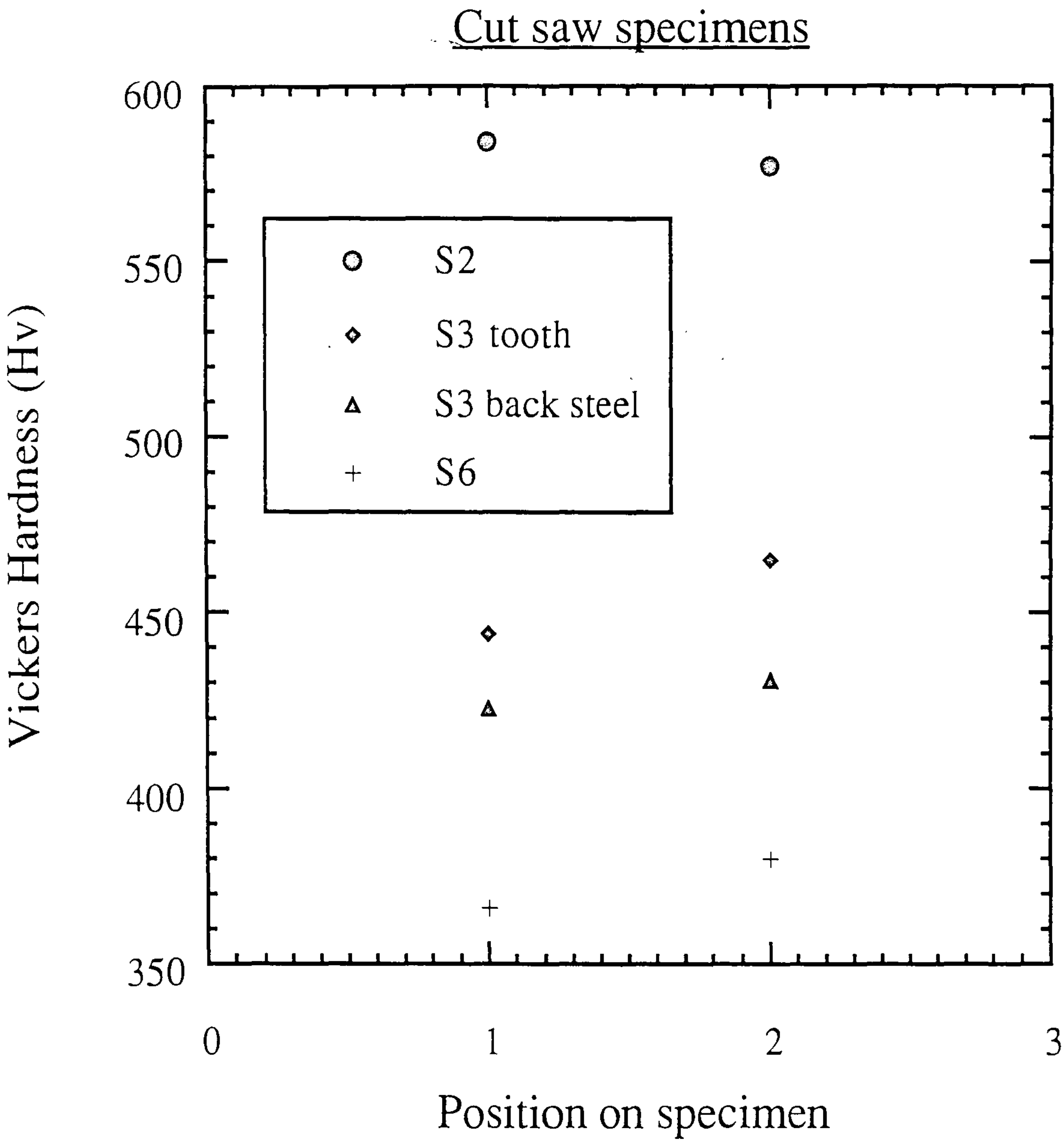


Figure 65 Hardness of Cut Saw Specimens

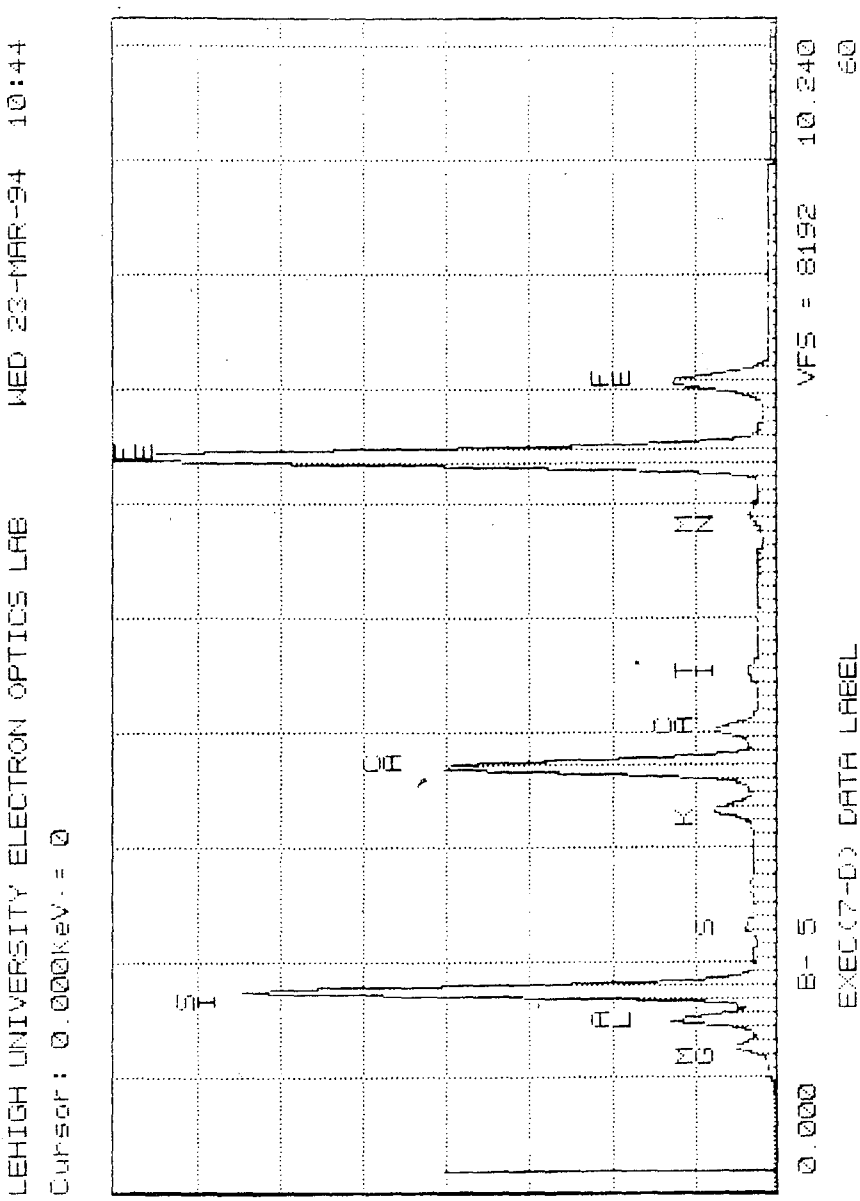


Figure 63 Spectra from slag in S3 (EDS)

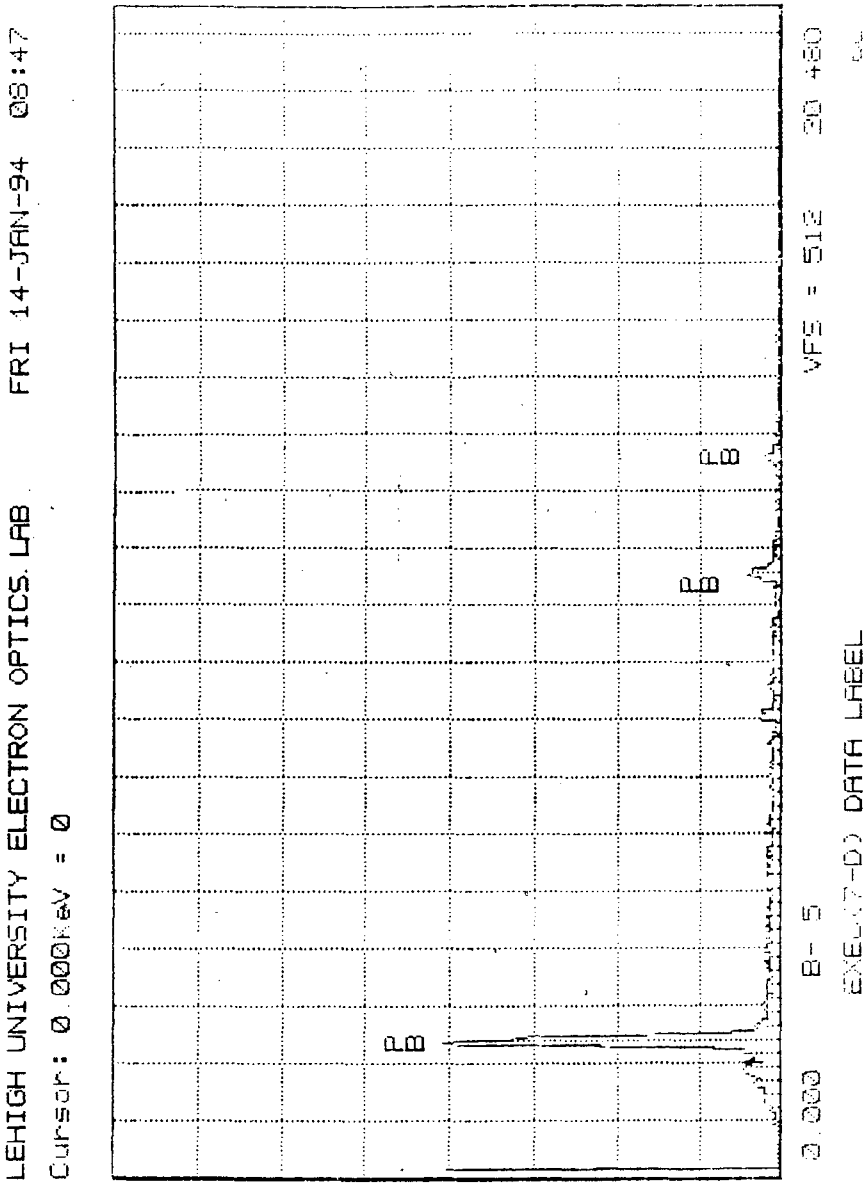


Figure 70 Spectra from Pb inclusions in S2 (EDS)

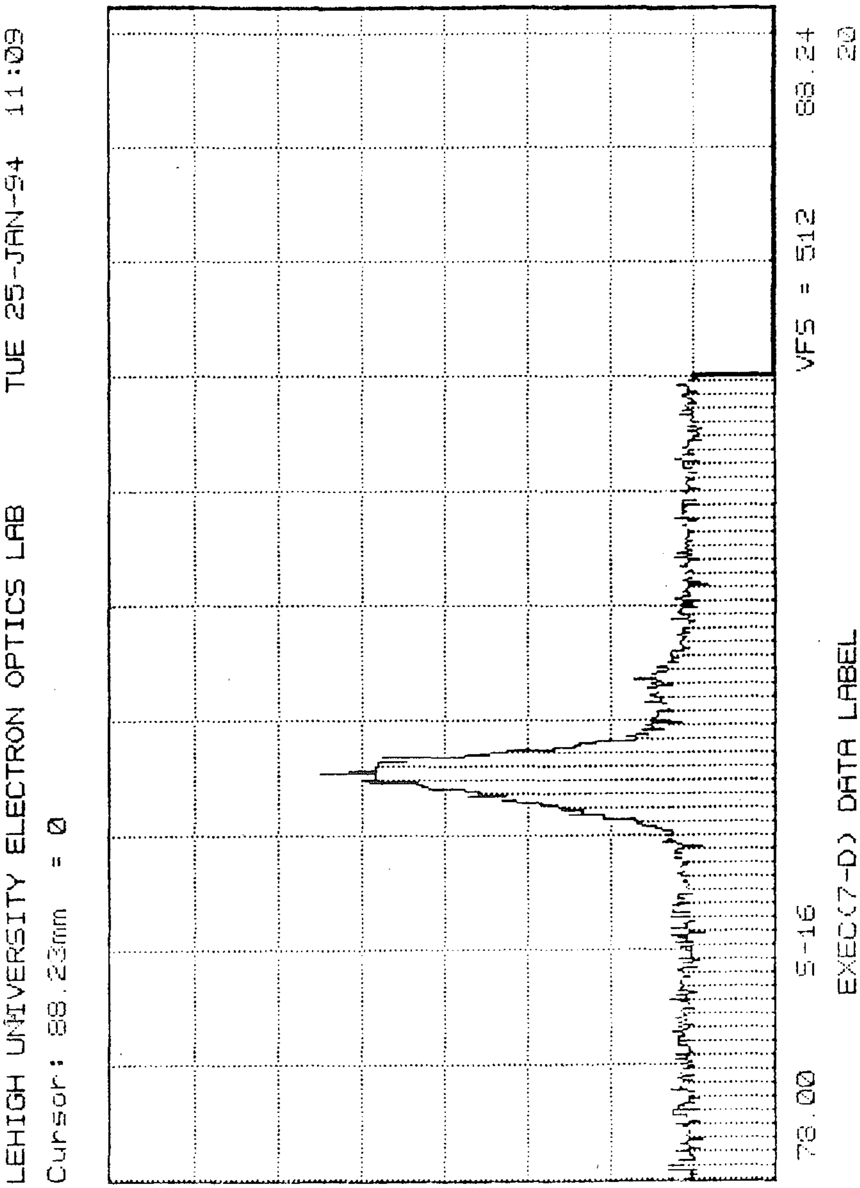


Figure 71 Spectra from Pb inclusions in S3 (WDS)

Table 3

Table for Plane Blades

sample	H-avg steel	H-avg iron	H-interface	origin	maker	structure-st	structure-ir	date	comments **
PB1	597	198	597.5	L	Cogdell, W.	M	F	1730	no oxide
PB2	609	149	608	L	Cogdell, J.	M	F	1750	no oxide
PB3	594	149	321.5	L	Madox	M	F	1748	oxide
PB4	608	165	651	L	Mutter	M	F	1766	no oxide
PB5	586	219	566	L	Mosely & Son	M	F	1810	oxide
PB6	596	153	139.5	L	Walton	LLM	F	pre 1785	oxide

legend

LLM = large lath martensite
M = martensite
F = ferrite
L = London
** refers to the presence of a thin oxide layer at the weld interface
note : all hardness measurements have the units Hv

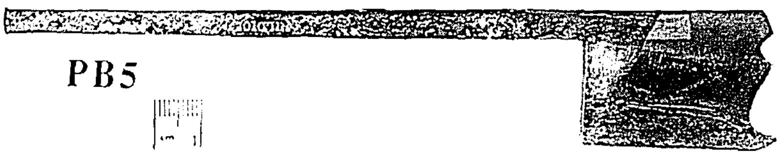


Figure 2

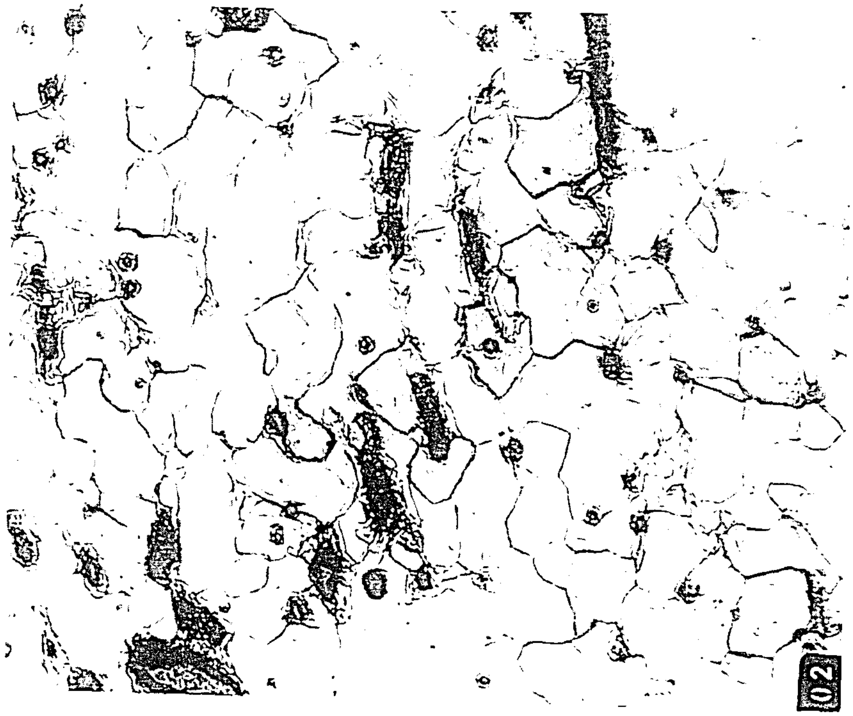


Figure 50 PB3, 320X, 2%nital

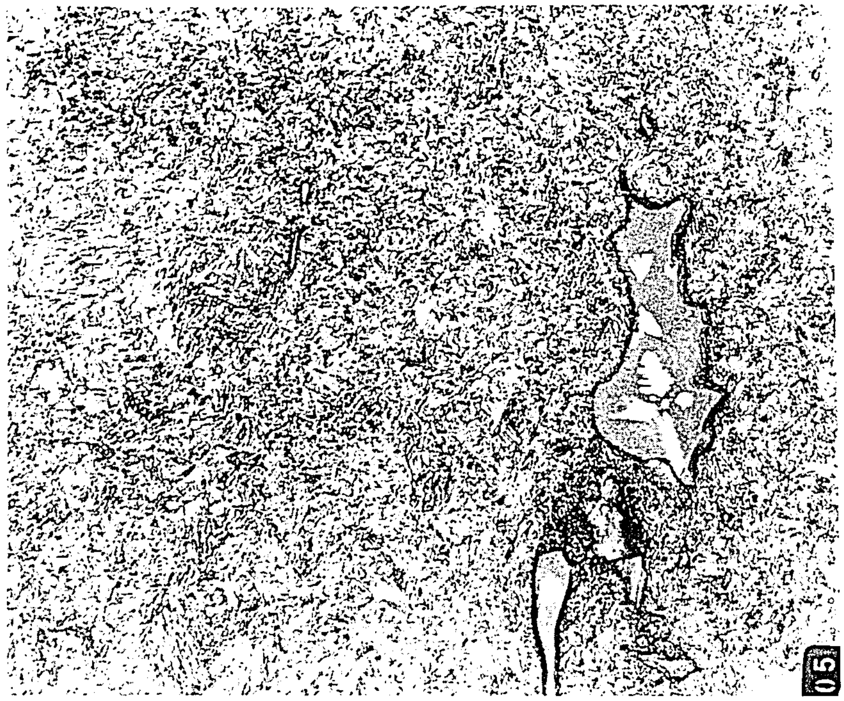


Figure 51 PB3, 500X, 2%nital

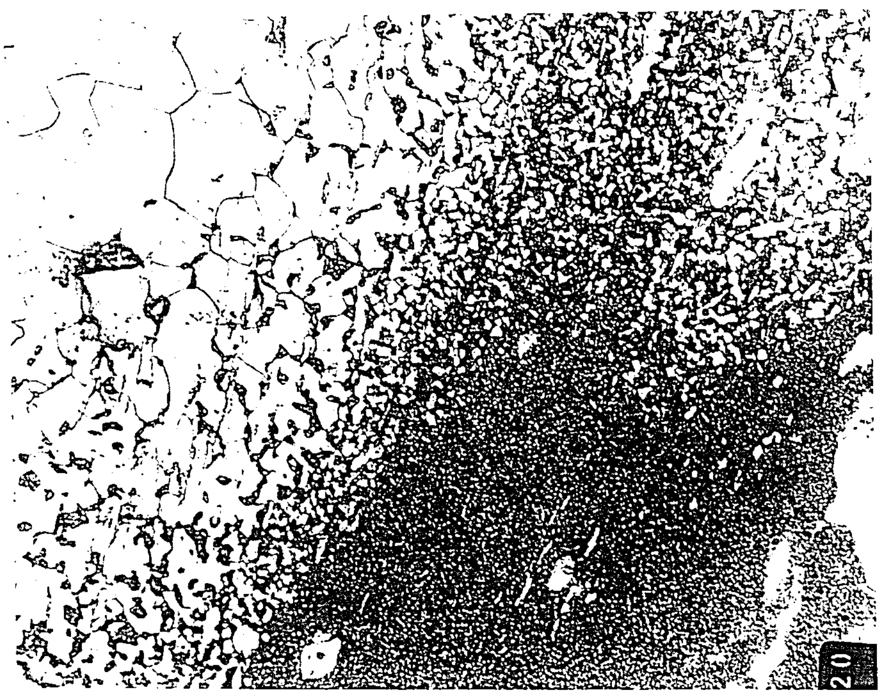


Figure 52 PB1, 160X, 2%nital

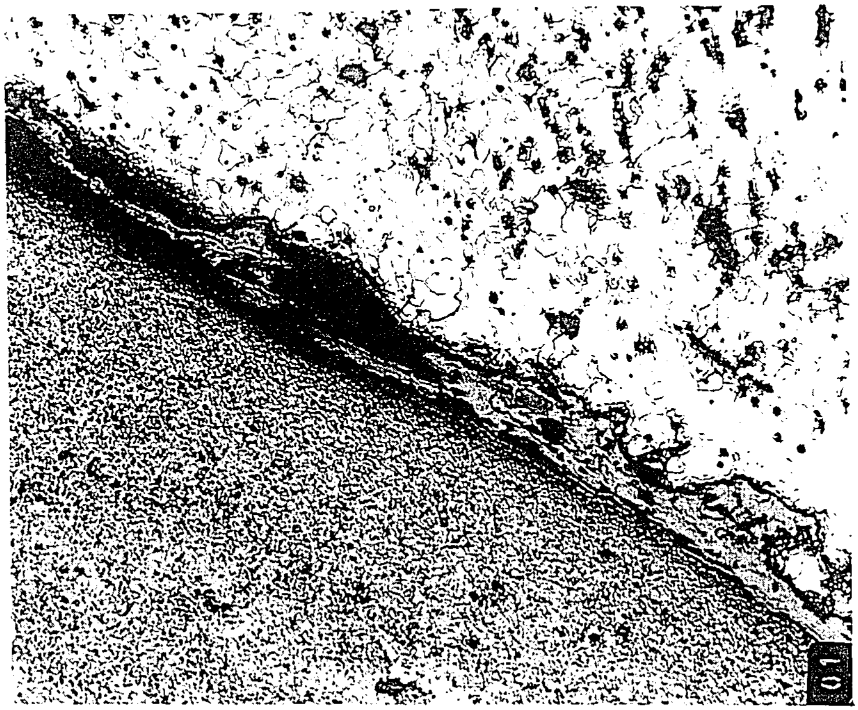


Figure 53 PB3, 100X, 2%nital

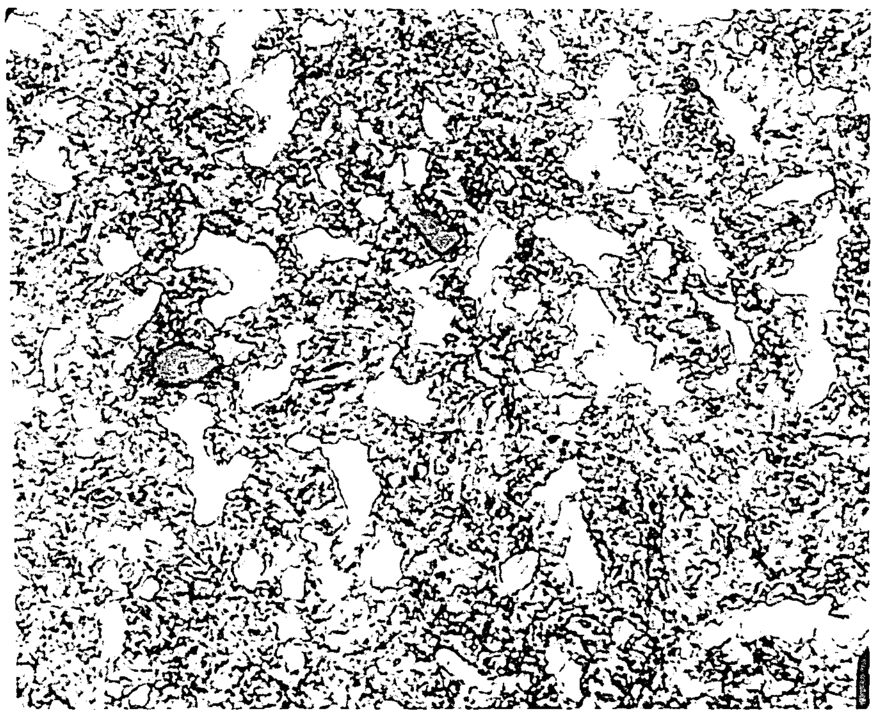


Figure 56, MPS3, 800X, 2%nital

Figure 58 MPS3, 800X, 4%picral

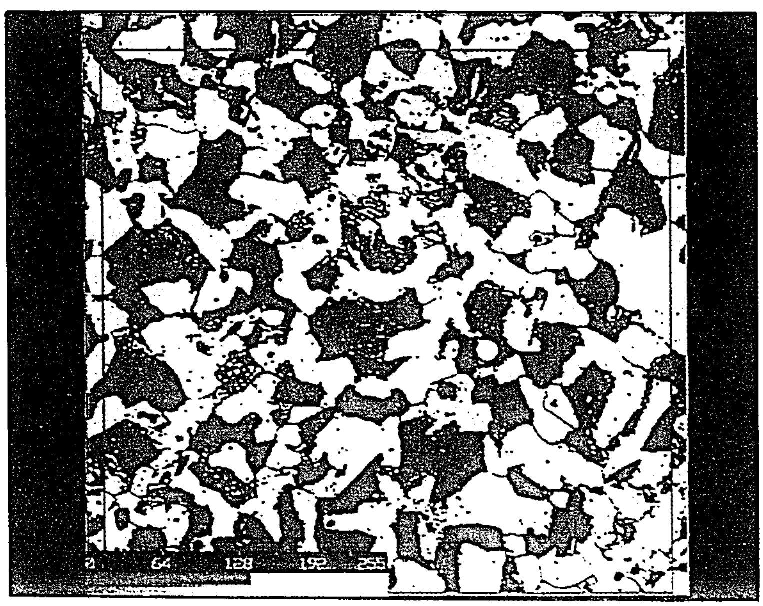


Figure 59 MPS3

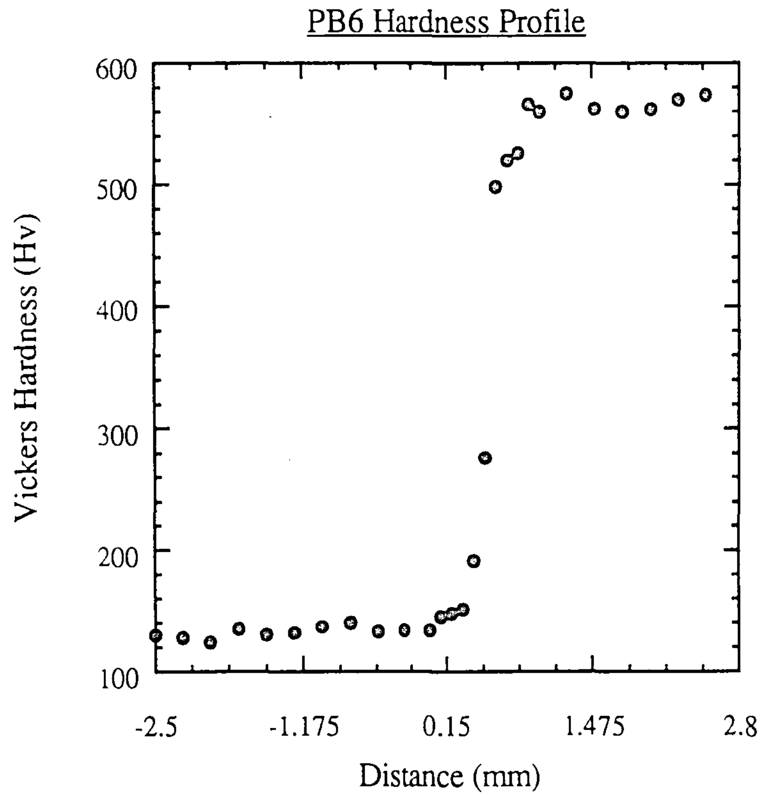


Figure 55 Hardness profile for PB6

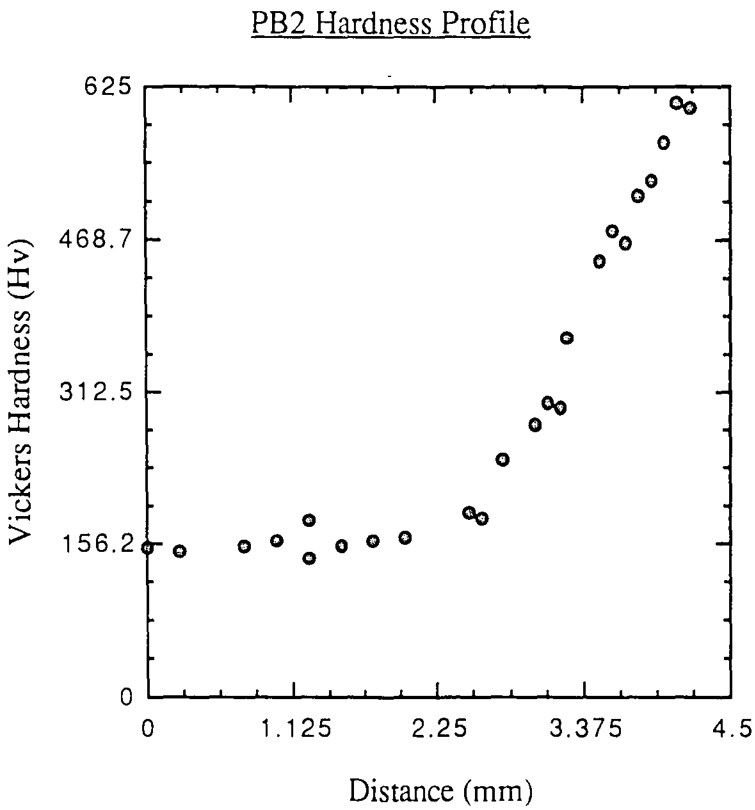


Figure 54 Hardness Profile for PB2

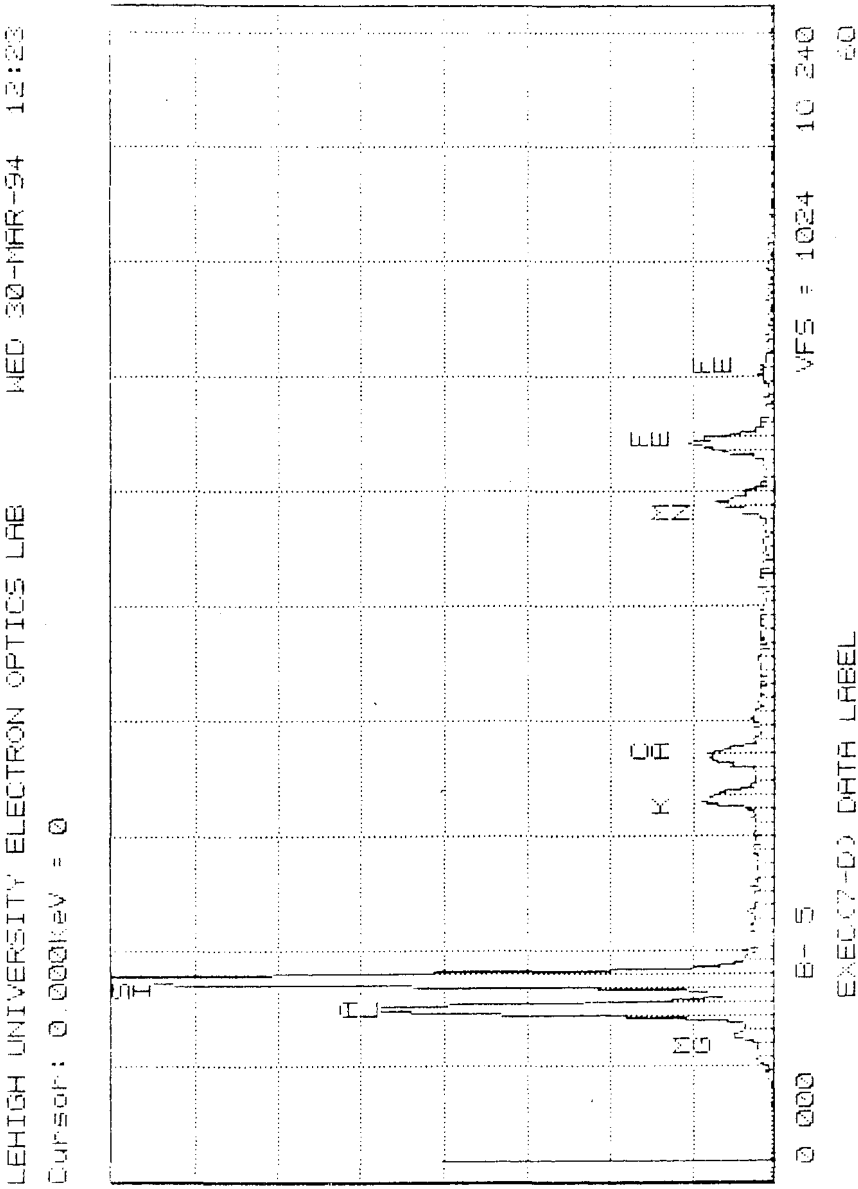


Figure 60 Spectra from MPS3 steel in slag (EDS)

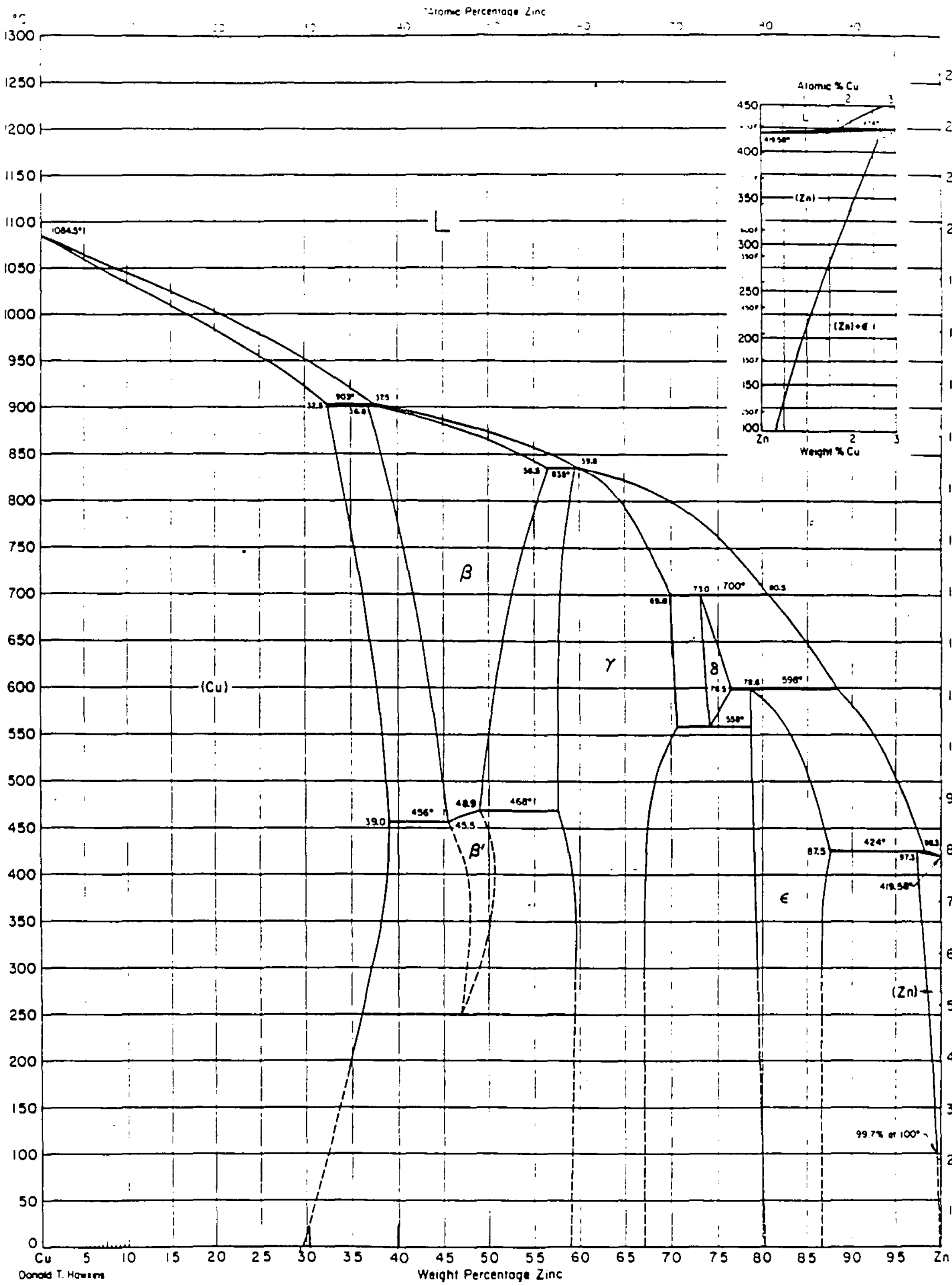


Figure 8 Cu-Zn phase diagram (ANO '73)

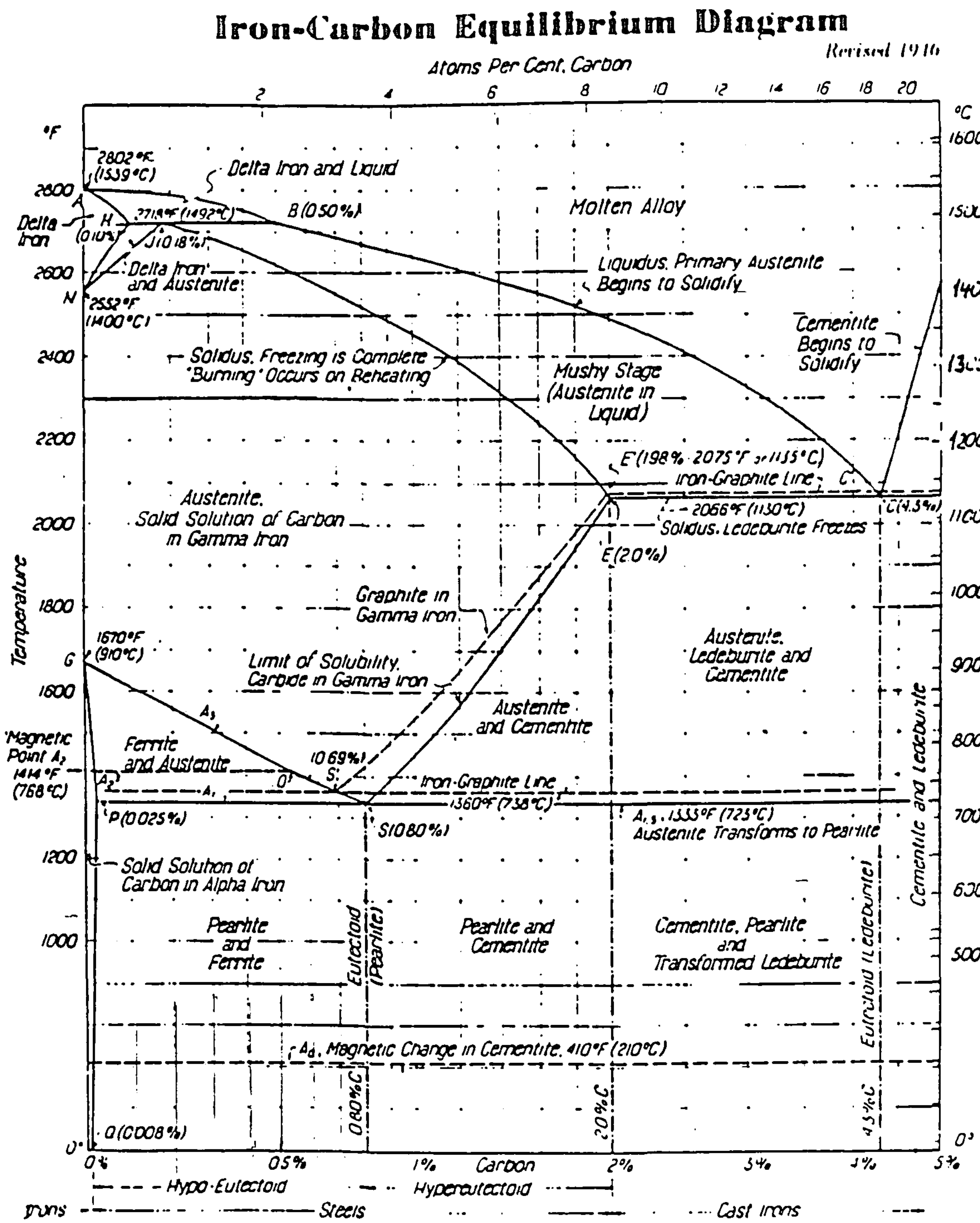


Figure 57 Fe-C phase diagram

SiO_2–P_2O_5

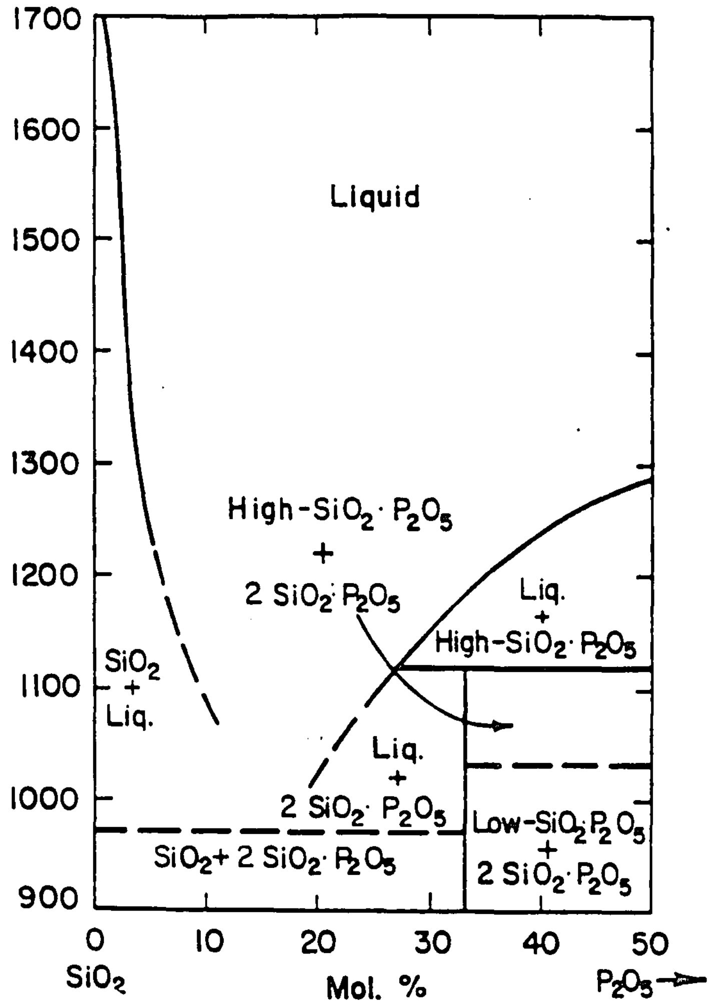


FIG. 364.—System SiO_2–$SiO_2 \cdot P_2O_5$.

T. Y. Tien and F. A. Hummel, *J. Am. Ceram Soc.*, 45 [9] 424 (1962).

Figure 104 SiO_2/P_2O_5 phase diagram (LEV '64)

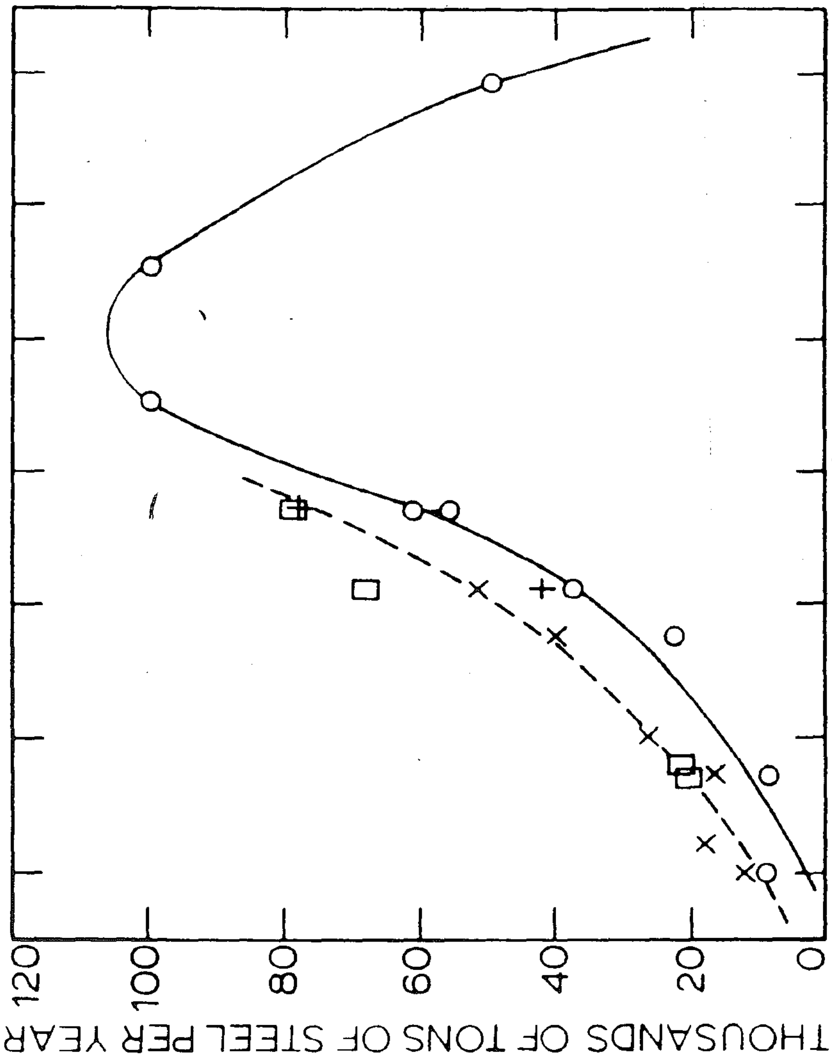


Figure 102 Useage of Crucible Steel (BAR '81b)

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Appendix A

Testing Objectives

Saws

1) To determine the general nature of saw blade materials, fabrication and hardening/tempering. (p.56)

2) To determine the nature of London Spring, German and Cast steels. (p.57)

3) To determine the materials from which the unmarked saws were made (and any suggestions of differences between London, Birmingham and Sheffield products). (p.56, 57)

Plane Blades

1) To determine the general nature of molding plane blades and to determine if there are changes in the blades' materials, fabrication or hardening/tempering over the period. (p. 57, 58)

Mortising Chisels

1) What was the nature of mid-century examples manufactured in Birmingham? (p.59)

2) Are there differences in these Birmingham chisels and early Sheffield examples? (p.59)

3) Does the range of Newbould chisels reflect changes in Sheffield products during the late eighteenth/early nineteenth centuries? (p.59)

4) How consistent are the chisels marketed by one manufacturer (using the Green tools)? (p.59)

5) How do the roughly contemporary Law and Newbould chisels compare to the Greens? (p.59)

6) How does the early nineteenth-century Law chisel compare to the somewhat earlier examples? (p.60)

7) What is the material of the un-specified Cam chisel? (p.60)

8) What is the nature of the Cast Steel in the so marked Cam chisel? (p.60)

9) What can we learn from the broken tang Weldon piece? (p.22, 23)

Socket and Tanged chisels

1) The general materials, fabrication and hardening/tempering of general purpose chisels. (p.60)

2) The nature of steel in mid-century Birmingham chisels. (See Table 5)

3) A comparison of Birmingham and Sheffield examples. (p.60)

4) A comparison of any differences in hardening/tempering between tanged (relatively light duty) and socket (heavy duty) chisels. (See table 5)

5) Whether "CS" refers to Cast Steel and whether the chisel so marked is made of solid cast steel rather than consisting of a steel/iron laminate. (p.61)

<u>Appendix B</u>

Short Guide to Finding Figures for Each Tool

S1	p.114, 117, 119
S2	p.114, 115, 119, 120, 122
S3	p.114, 116, 117, 118, 119, 120, 121, 123
S4	p.114, 116, 119
S5	p.114, 119
S6	p.114, 115, 119, 120
PB1	p.124, 126
PB2	p.124, 130
PB3	p.124, 125, 126, 127
PB4	p. 124
PB5	p.124, 125
PB6	p.124, 129
MPS	p.124, 127, 128, 131
SC1	p.91, 96, 97, 98, 99, 100, 101, 102, 104, 105, 106, 107, 108, 111, 112, 113
SC2	p.91, 92, 103, 107
TC1	p.91, 92, 107
TC3	p.91, 93, 94, 95, 107
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Vita for Donna L. Belcher

Donna Belcher was born in 1970 to Don and Grace Belcher. She later graduated from Purdue University in 1992 with a BS in materials science and engineering. She came to Lehigh University in the fall of 1992 to study archaeometallurgy with Dr. Mike Notis.

END

OF

TITLE